

Analysis of Dynamic Failure of High-Density Polyethylene (HDPE) as Engineering Materials

ABSTRACT

Advances in engineering reliability and variability have allowed for a vast investigation into the dynamic failure of engineering materials in recent times than previously possible. This report aims to investigate and review the basic model in dynamic failure of High-Density Polyethylene (HDPE) engineering materials by fatigue, through the dynamic crack initiation and growth as in brittle materials, ductile materials and elastic-plastic solids as in layered materials and composites and adiabatic shear bending in ductile materials. Slow crack growth (SCG) under sustained loads (pressure and axial loads) is one of the limiting failure modes that affect the long term performance of High-Density Polyethylene (HDPE) pressure material identified for use in replacement of existing steel material. This report also compares the resistance to the SCG exhibited by the parent and fusion HDPE materials in the Single Edge Notch Tension (SENT) specimen testing. Analysis of the crack growth resistance parameter through crack-mouth-opening-displacement (CMOD), and crack-opening-angle (COA) revealed a marked difference between the parent and fusion HDPE material.

Keywords: Dynamic failure, High-Density Polyethylene (HDPE), dynamic crack initiation, Crack-Mouth-Opening-Displacement (CMOD), and Crack-Opening-Angle (COA)

1.1 INTRODUCTION

Failure has been a serious problem in the use of materials since the beginning of recorded history. These catastrophic failures were a driving force for the development of material science and engineering [3]. Failure can be described as any change of properties which make the material or component functionally, structurally or aesthetically unacceptable [5]. In the last few decades, engineering polymers have succeeded in replacing metals in many demanding applications and such failures will become even more important. It is often necessary to understand why polymer failure has occurred so that measures can be taken to prevent its reoccurrence [4]. Polymeric materials are sensitive to processing and affected by the environment, time and temperature during storage, transportation and service. Especially the long-term properties are frequently “unpredictable” [3].

Failure is defined as the separation of a body into two or more parts by tension or compression [1; 3]. Dynamic failure of engineering materials (ductile metals) in tension takes place by the nucleation, growth and coalescence of voids [1]. For brittle materials, it takes place by the nucleation, growth, and coalescence of cracks [1]. As the rate at which materials are deformed

41 increases, the following effects play an increasingly important role: (a) Mass inertia which leads
42 to the propagation of elastic, plastic, and shock waves. (b) Thermal inertia, which is the thermal
43 diffusion distance decreases as the time for deformation decreases, leading to pronounced
44 temperature in homogeneities within the material. (c) Thermal activation and viscosity, which is
45 the response for dislocations (the primary carriers of plastic propagation velocities phonon and
46 electron viscosity may determine the response [3]. [4]. these three effects determine the elastic,
47 plastic and failure response of materials. The infinite complexity of the morphological
48 characteristics of failure can be rationalized by the interplay of the above-named effects (mass
49 inertia, thermal inertia, thermal activation and viscosity) with microstructural characteristics of
50 materials [1]. Dynamic failures of High-Density Polyethylene (HDPE) engineering materials can
51 be classified into three groups [2]:

52 **(a) Tensile failure;** which is the state of stress and the dynamics of generation, propagation, and
53 interconnection of flaws dictates the morphology. Under uni-axial strain conditions, this failure
54 is called "spalling" [3]. **(b) Compressive failure;** under compressive traction localized regions
55 of tension can be generated in the microstructure, which gives rise to failure. Although metals
56 (gold, silver, etc.) are mostly immune to this type of failure, less ductile metals (eg., tungsten,
57 steel), ceramics, composites are subjected to this type of failure [2]. **(c) Shear failure of shear**
58 **localization,** with usually micro-structural or thermal origins, often leads to failures. It should be
59 emphasized that the adiabatic shear band is the precursor event and that it provides a path for
60 crack propagation (fragilized or softened material) which is a tensile stress phenomenon. Failure
61 of materials may have huge costs. Causes included improper materials selection or processing,
62 the improper design of components, and improper use.

63
64 Failure in polymer components can occur at relatively low stress levels (far below the tensile
65 strength in many cases) due to long-term stress (creep-rupture), cyclic stresses (fatigue failure) or
66 liquid agents (environmental stress cracking) [5]. When a polymer is stressed in the air to just
67 below its yield point, stress cracking can occur after some time. However, when simultaneously
68 exposed to both stress and a chemical medium, this will result in a sharp reduction of the time to
69 failure [7]. This type of failure has been named environmental stress cracking (ESC). ESC has
70 been a subject of extensive investigations for almost 50 years [5]. It has deserved much attention
71 because approximately 15 to 20 % of all failures of plastic components in service are due to ESC
72 [13].

73
74 At this point, the industry was confronted with numerous reports of polyethene failure.
75 Polyethene was reported to be unsatisfying for cable usage, and it was found to crack violently
76 on contact with methanol at room temperature. Therefore, the problem of ESC is very important
77 for many applications including packaging industry (bottles, containers, foils, films, etc.),
78 electric industry and electronics (wire and cable insulation), medicine (labware, caps, implant
79 components, etc.), automobile industry (tanks, pipes, coatings, etc.) and many more [1].

80

81 Therefore, Dynamic Failure of High-Density Polyethylene (HDPE) engineering material is a
 82 form of (silent) failure where the material separates in pieces due to stress at temperatures below
 83 the melting point [5]. The failure is termed ductile or brittle depending on whether the elongation
 84 is large or small. Steps in failure (response to stress) are Track formation and Track propagation
 85 [11]. A common characteristic of these failure phenomena is the rapid loss of stress carrying
 86 capability in time scales as depicted in table 1 [10; 16].

87 **Table 1: Ductile vs. brittle fracture**

	Ductile	Brittle
Deformation	Extensive	Little
Track propagation	Slow, needs stress	Fast
Type of materials	Most metals (not too cold)	Lower
Warning	Permanent elongation	None
Strain energy	Higher	Lower
Fractured surface	Rough	Smoother
Necking	Yes	No

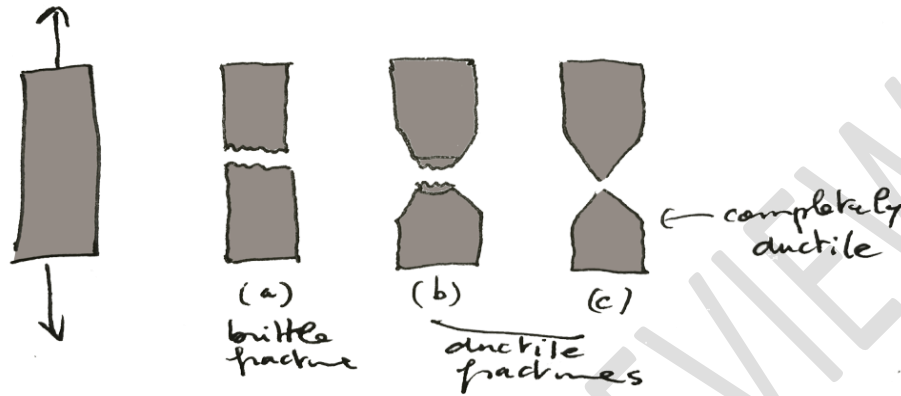
88 **1.1.1 Ductile Fracture:** In High-Density Polyethylene (HDPE) material, the material fracture
 89 slowly and deform plastically before seeing a sign of a crack that is because those parts are built
 90 in a ductile material as in Figure 1 [18]. Stages of ductile fracture are; Initial necking, small
 91 cavity formation (microvoids), Void growth (ellipsoid) by coalescence into a crack, Fast crack
 92 propagation around the neck. Shear strain at 45° final shear fracture (cup and cone), The interior
 93 surface is fibrous, irregular, which signify plastic deformation [16].



94

95 **Figure 1: Depicts the failure of ductile materials**

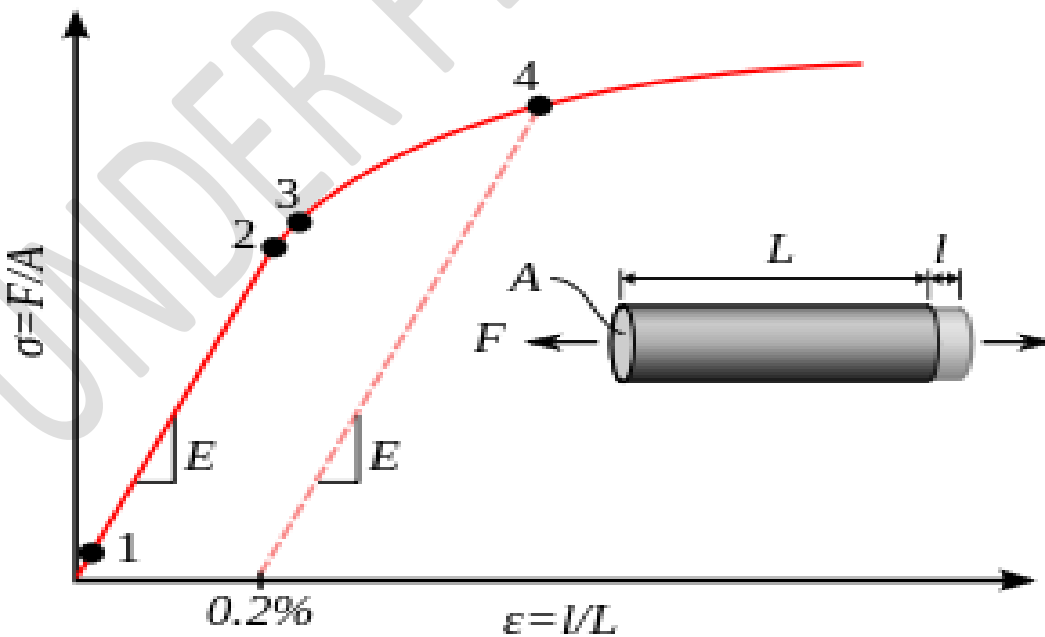
96 Ductile fracture with characteristic distortion and shear lip, steel is a ductile material. It extends
 97 and deforms before failing as in figures 2b and 2c below [16]. That is represented in the stress-
 98 strain curve as in figure (3). These figures 2a, 2b, 2c depicts the difference between brittle,
 99 ductile and semi-ductile materials respectively:



100

101 **Figure 2: Depicts the difference between brittle, ductile and semi-ductile materials**

102 **1.1.2 Brittle Fracture:** in this failure, there is no appreciable deformation and crack propagation
 103 is very fast as in figure 2a. In most brittle materials, crack propagation (by bond breaking) is
 104 along specific crystallographic planes (*cleavage* planes) [10]. This type of fracture is
 105 transgranular (through grains) producing grainy texture (or faceted texture) when cleavage
 106 direction changes from grain to grain. In some materials, the fracture is intergranular [16].



107

108 **Figure 3: Typical yield behaviour of engineering materials**

109 **Offset yield strength:** It is often difficult to precisely define yielding due to the wide variety of
110 stress-strain curves exhibited by real materials. Also, there are several possible ways to define
111 yielding:

112 **True elastic limit:** The lowest stress at which dislocations move. This definition is rarely used
113 since dislocations move at very low stresses, and detecting such movement is very difficult.

114 **Proportionality limit:** Up to this amount of stress, stress is proportional to strain (Hooke's law),
115 so the stress-strain graph is a straight line, and the gradient will be equal to the elastic modulus of
116 the material.

117 **Elastic limit (yield strength):** Beyond the elastic limit, permanent deformation will occur. The
118 elastic limit is, therefore, the lowest stress point at which permanent deformation can be
119 measured. This requires a manual load-unload procedure, and the accuracy is critically
120 dependent on the equipment used and operator skill. For elastomers, such as rubber, the elastic
121 limit is much larger than the proportionality limit. Also, precise strain measurements have shown
122 that plastic strain begins at low stresses [16].

123 **Yield point:** The point in the stress-strain curve at which the curve levels off and plastic
124 deformation begins to occur.

125 **Offset yield point (proof stress):** When a yield point is not easily defined based on the shape of
126 the stress-strain curve an *offset yield point* is arbitrarily defined. The value for this is commonly
127 set at 0.1 or 0.2% plastic strain. The offset value is given as a subscript, e.g., $R_{p0.2}=310$ MPa.
128 High strength steel and aluminium alloys do not exhibit a yield point, so this offset yield point is
129 used on these materials.

130 **Upper and lower yield points:** Some materials, reach an upper yield point before dropping
131 rapidly to a lower yield point. The material response is linear up until the upper yield point, but
132 the lower yield point is used in structural engineering as a conservative value. If the material is
133 only stressed to the upper yield point, and beyond, Lüders bands can develop [16]. When a
134 propagating crack tip passes a material point, the material point instantaneously separates into at
135 least two parts. In the ordinary numerical models with nodal release techniques, this sudden
136 unloading process often produces spurious oscillations. The crack propagation velocity along a
137 bimaterial interface can become extremely fast and can exceed the shear wave velocity of the
138 compliant material [18].

139 **1.1.3 Crack Initiation and Propagation:** Stages in fatigue failure are categorized as follows:
140 (I). Crack initiation at high-stress points (stress raisers). (II). Propagation (incremental in each
141 cycle). (III). Final failure by fracture [4]

142 Stage I - propagation

- 143
- Slow
 - Along crystallographic planes of high shear stress
 - Flat and featureless fatigue surface
- 145

146 Stage II - propagation

- 147
- Crack propagates by repetitive plastic blunting and sharpening of the crack tip [7].
 - Crack Propagation Rate (not covered)
- 148

149

150 **2.1 HOW STRESS CRACKING OCCURS IN ENGINEERING MATERIAL**

151 The failure of a container to resist ESC can be the result of stored stresses acquired in the
152 moulding or extrusion operation [6]. Stress cracking agents, such as the liquids mentioned above,
153 migrate into minute cracks in the crystalline areas of the polyethene molecules forming the
154 surface of the container. These microscopic cracks are a result of a breakdown of the polymer
155 chains in the case of acids and solvents, and a “wetting out” of the surface in the case of
156 detergents, largely due to their surfactant components [5].

157

158 In either case, the surface tension between the crystalline layers is reduced. What happens is that
159 once a microscopic surface imperfection propagates or "zippers open" to a full-fledged break in
160 the bottle. ESC failures are accelerated by high temperatures and additional external stresses
161 such as top-loaded storage [9].

162

163 **2.2 SIGNIFICANCE OF ENVIRONMENTAL STRESS CRACKING RESISTANCE**

164 **2.2.1 The Stress Factor**

165 As the name suggests, stress cracking requires the polymer to have exposure to intrinsic residual
166 stress or an externally applied stress. If the plastic moulding is completely free of stress, then no
167 stress cracking will occur [5]. Even polymers exposed to liquids or vapours that have a swelling
168 or wetting effect will not undergo ESC unless there is an externally applied or moulded-in stress
169 present. External stress may be the result of component assembly (composite formulation),
170 improper packing or storage, incorrect use, etc [8].

171

172 The definition of cracking due to stress is defined in many standards. It is stated as the internal or
173 an external crack in the plastic caused by stresses less than its short-term mechanical potency
174 [20]. This type of cracking usually consists of brittle cracking with no ductile or little drawing of
175 plastic material from the adjoining surface failure. Slow growth in cracks is another term used to
176 explain stress cracking [15].

177

178 Environmental stress cracking (ESC) in plastics means the failure at about room temperature due
179 to continuously acting external and/or internal stresses in the presence of surface-active
180 substances (known as stress cracking agents) such as alcohols, soaps, dyes, agents containing
181 moisture [12]. Although ESC results from the interaction of the polymer with certain chemicals,

182 it is not a chemical reaction between the polymer and the active environment [17]. The stress
183 cracking agents do not cause any chemical degradation of the polymer but they accelerate the
184 process of macroscopic brittle-crack formation

185

186 **3. REVIEW OF RELATED LITERATURE**

187 [19; 20], used a modified ENF specimen to determine the mode II dominated dynamic
188 delamination fracture toughness of fibre composites at high crack propagation speeds. A strip of
189 adhesive film with higher toughness was placed at the tip of interlaminar crack created during
190 laminate lay-up [7]. The objective was to delay the onset of crack extension and produce crack
191 propagation at high speeds (700 m/s). Sixteen pure aluminium conductive lines were put on the
192 specimen edge side using the vapour deposition technique, to carry out crack speed
193 measurements [9]. The authors concluded that the mode II dynamic energy release rate of
194 unidirectional S2/8553 glass/epoxy composite seems to be insensitive to crack speed within the
195 range of 350 and 700 m/s. The authors also simulated mixed mode crack propagation by moving
196 the pre-crack from the mid-plane to 1/3 of the ENF specimen thickness of unidirectional
197 AS4/3501-6 carbon/epoxy laminates [9; 15].

198 The majority of the experimental studies consider unidirectional laminates. [5; 6; 19; 20],
199 performed an experimental investigation of dynamic crack initiation and growth in unidirectional
200 fibre-reinforced polymeric-matrix thick composite plates. Edge-notched plates were impacted in
201 a one-point bend configuration using a drop-weight tower. Using an optical method the authors
202 carried out a real-time visualization of dynamic fracture initiation and growth for crack speeds up
203 to 900 m/s. They verified that the elastic constants of the used material are rate sensitive and the
204 measured fracture toughness values are close to those types of epoxies. This was considered
205 consistent because in unidirectional lay-ups crack initiation and growth occurs in the matrix.

206 [7; 12], have suggested that the dynamic fracture behaviour of materials depends on the balance
207 between the energy released by the structure over a unit area of crack propagation (G) and the
208 material resistance (R), which can be viewed as the energy dissipated in creating the fracture
209 surface. When unstable crack growth occurs, the difference $G-R$ is converted into kinetic energy.
210 If G increases with crack growth the crack speed also increases because more energy is available.
211 The crack arrest will occur when G becomes lower than R and, consequently, no kinetic energy
212 is available for crack growth. Thus, it can be affirmed that fracture stability depends on the
213 variations of the strain energy release rate and the resistance of the material during crack growth

214

215

216 **4. STUDY BACKGROUND**

217 **4.1 Fatigue Failure Propagation**

218 Many materials will fail at lower stress when subjected to cyclic or repetitive loads than when
219 under static loads [2]. And it has been found experimentally that when a material is subjected to

220 repeated stress; it fails at stresses below the yield point stresses. This type of failure as regards
 221 engineering materials is known as fatigue [1]. Fatigue failure is caused through progressive crack
 222 formation which is usually fine and of microscopic size, which occurs even without any prior
 223 indication. For thermoplastics pipe materials, fatigue is only relevant where a large number of
 224 stress cycles are anticipated [5]. The important factors to consider are the magnitude of the stress
 225 fluctuation and the loading frequency. Where large stress fluctuations are predicted, fatigue
 226 design may be required where the total number of cycles in the operational lifetime of the pipe
 227 exceeds 100,000. For smaller stress cycles, a larger number of cycles can be tolerated [6].

228

229 **4.2 Modeling of Slow Crack Growth (SCG) Resistance of HDPE Parent Material**

230

231 In engineering material, Slow crack growth (SCG) under sustained loads (pressure and axial
 232 loads) is defined as the limiting failure process that affects the long term performance of High-
 233 Density Polyethylene (HDPE) pressure materials listed for use in replacement for existing one.

234

235 The Brown model employs a power-law function of SIF (KI), an exponential function of
 236 temperature (akin to an Arrhenius equation [5; 21]), and functions to capture geometric
 237 dimensions/constraint factors. The model, as shown in equation (1) for life prediction bases its
 238 approach on the PENT failure time and has been used extensively by the plastic pipe industry
 239 and has recently been considered by the nuclear industry [5; 17; 20].

240

$$241 \quad tf = [tPENT] \left(\frac{0.468}{KI}\right)^n \text{Exp}\left[\frac{Q}{R}\left(\frac{1}{T} - \frac{1}{353}\right)\right] \quad 1$$

242

where tf is time to failure (service life)

243

$tPENT$ = PENT failure time

244

KI = stress intensity factor (=0.468 MPa-m^{0.5} for PENT specimen)

245

n = stress intensity exponent (~2.5 to 4.5)

246

Q is resin activation energy (~85 to 110 kJ/mol)

247

R is universal gas constant 0.008314 kJ/mol/°K, and T is absolute temperature in °K (=353 °K for
 248 PENT specimen)

249

The Brown model was calibrated using the slope $n=3.25$, PENT value $tPENT=60$ hours, and the
 250 temperature shift function, $Q=100$ from test data for the HDPE PE4710 bimodal butt-fusion joint
 251 materials [5; 13; 20; 21].

252

253 **4.3 Experimental Investigation on the SCG Resistance of HDPE Parent Material**

254 The service life prediction models employ the failure time from a PENT, SENT [19] test to
 255 obtain the performance life of HDPE material with a detected flaw depth, and operating
 256 temperature, and pressure [5; 12; 13]. Recent work indicate that the PENT and SENT specimens
 257 to have a SIF and constraint/transverse T -stress (coefficient $\beta = [T(\pi\alpha)^{0.5}]/KI$) that is close to
 258 that of a surface crack in an HDPE pipe and have been the focus of the investigations [5; 12; 14;
 259 18].

260 To determine the variability of failure times and validate service model predictions; tests were
 261 conducted at PENT $KI = 14.942 \text{ MPa}\cdot\text{mm}^{0.5}$ (430 $\text{psi}\cdot\text{inch}^{0.5}$), SENT tests at $KI = 18.486 \text{ MPa}\cdot\text{mm}^{0.5}$
 262 $\text{mm}^{0.5}$ (532 $\text{psi}\cdot\text{inch}^{0.5}$), and at $KI = 10.181 \text{ MPa}\cdot\text{mm}^{0.5}$ (293 $\text{psi}\cdot\text{inch}^{0.5}$) [5; 13; 21].

263 Comparison of parent material SCG failure time (7893 hours when tested at 95°C and KI
 264 $=18.486 \text{ MPa}\cdot\text{mm}^{0.5}$ (532 $\text{psi}\cdot\text{inch}^{0.5}$)) with the butt-fusion joint material SCG failure time (less
 265 than 100 hours when tested at 95°C and $KI = 18.486 \text{ MPa}\cdot\text{mm}^{0.5}$ (532 $\text{psi}\cdot\text{inch}^{0.5}$)) were discussed
 266 in [5; 6; 18]. Table 1 shows the list of butt-fusion joint SENT and PENT specimens tested from
 267 the 10k and 2k PE4710 HDPE resins and their corresponding failure times were evaluated as in
 268 equation 1[5; 18;19].

269 **Table 2: SCG Tests on Butt-fusion Joint (B) Materials of PE4710 Bimodal HDPE 10,000**
 270 **hours PENT (10k) and 2,000 hours PENT (2k) Resins [5; 18]**
 271

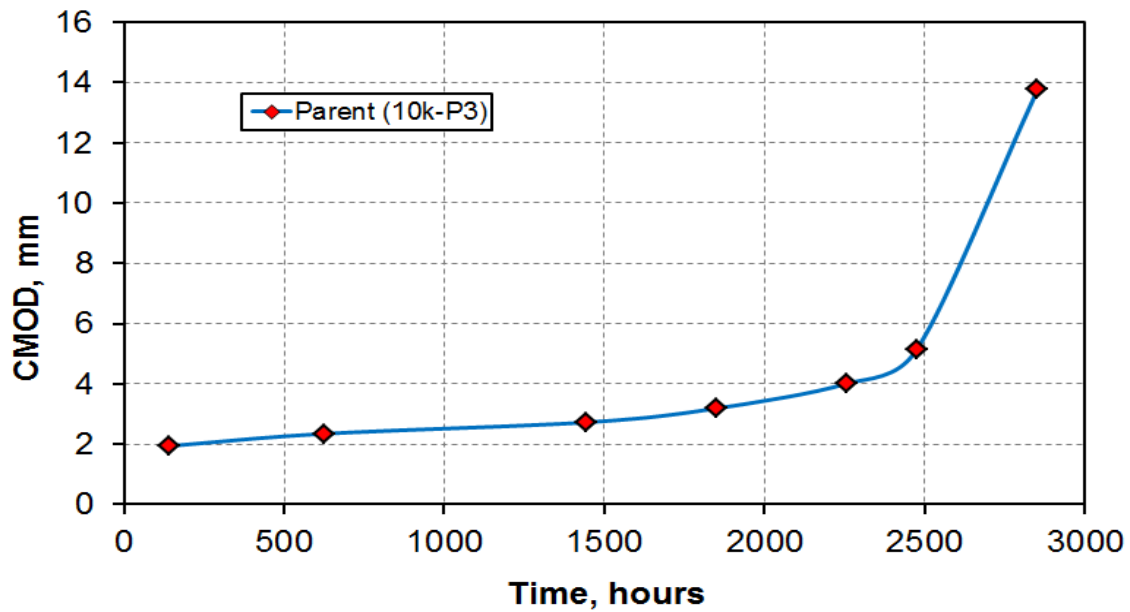
Specimen Name	Test Frame	SIF, KI $\text{MPa}\cdot\text{mm}^{0.5}$ ($\text{psi}\cdot\text{inch}^{0.5}$)	Failure Time, hours	Test Temp. T, °C
10k –B3	L Creep	18.486 (532)	20	95
10k –B4	L Creep	18.486 (532)	67	95
10k –B5	L Creep	10.181 (293)	482	95
10k –B11	PENT	14.942 (430)	8.8	95
10k –B12	PENT	14.942 (430)	6	95
10k –B7	PENT	6.984 (201)	134	95
10k –B8	PENT	6.984 (201)	213	95
2k –B5	L Creep	19.112 (550)	73.9	95
2k –B7	PENT	14.942 (430)	10.3	95
2k –B8	PENT	14.942 (430)	19.3	95

272 The four stages of SCG in the 10k-P3 parent HDPE specimen that was correlated to the
 273 observations depicts initial crazing, accumulation of damage, the start of SCG, and final failure
 274 [5; 12;19;21]. The experimental data obtained during the crack growth over several hours of
 275 creep test time using equation 1 to model the crack-mouth-opening displacement (CMOD) and
 276 crack-opening angle (COA) as a function of time. The crack length at various times during the
 277 creep test was obtained by considering the crazed region and fibril breakage in the vicinity of the
 278 current crack front as seen on the surface of the HDPE SENT specimen [5; 12; 19] and the crack
 279 length measurements in these tests were chosen to be until the end of the crazing zone.

280 5. RESULT AND DISCUSSION

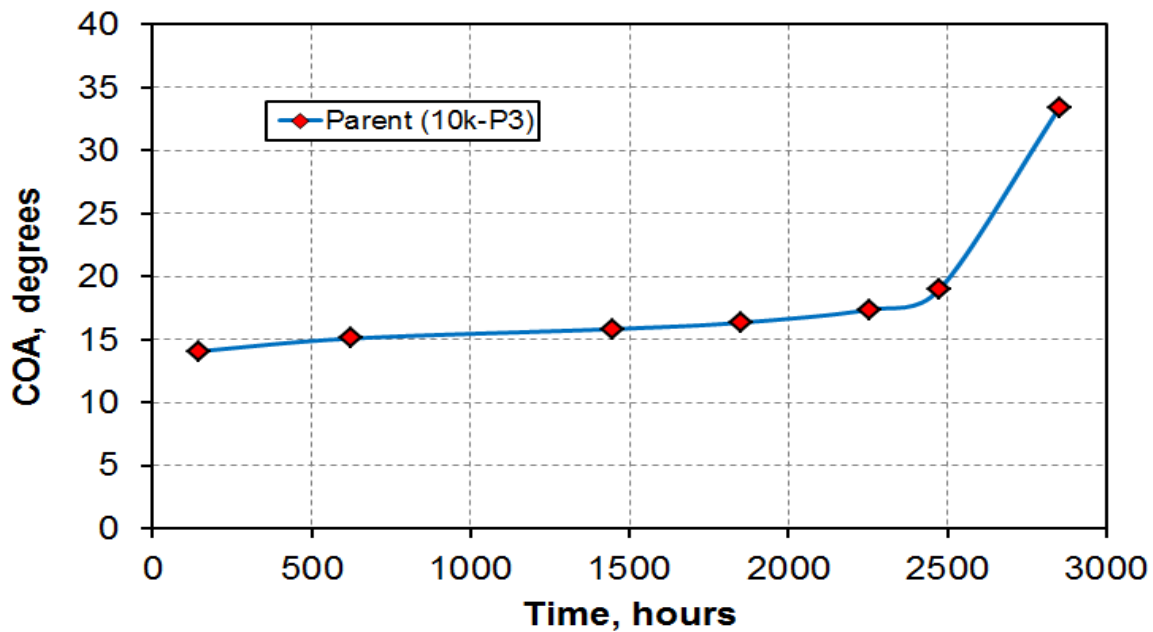
281 The failure process during slow-crack growth (SCG) in HDPE materials typically occurs as
 282 illustrated in Figures 4 to 6 were analyzed to determine the CMOD, CTOA, and crack length
 283 variations with time for the SENT specimen 10k-P3 with the crack in the parent HDPE material.
 284 Figure 4 illustrates the variation of the crack-mouth-opening displacement (CMOD) with time
 285 which is indicative of the toughness of the parent HDPE material. Similarly, Figure 5 also

286 illustrates the crack-opening angle (COA) with time which is also indicative of the toughness of
287 the parent HDPE material. While figure 6, illustrates the variation of the crack length with time
288 as the SCG occurred in the parent HDPE SENT specimen.



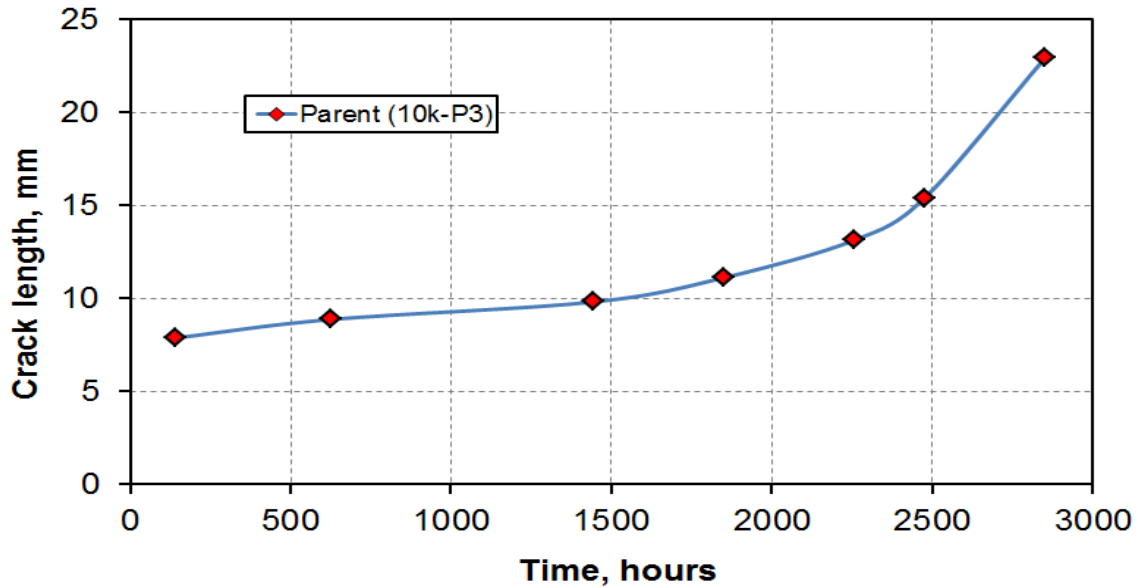
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290 **Figure 4: CMOD versus time during SCG in parent HDPE SENT specimen, 10k-P3**



291

292 **Figure 5: COA versus time during SCG in parent HDPE SENT specimen, 10k-P3**



293

294 **Figure 6: Crack length versus time during SCG in parent HDPE SENT specimen, 10k-P3**

295 Failure at low loads is in the elastic strain regime, requires a large number of cycles (type. 10^4 to
 296 10^5). At high loads (plastic regime), one has low-cycle fatigue ($N < 10^4 - 10^5$ cycles).

297 **6. CONCLUSIONS**

298 The following conclusions were drawn from this work:

- 299 • Observations of SCG at both higher ($\sim 18.486 \text{ MPa}\cdot\text{mm}^{0.5}$ (532 psi-inch $^{0.5}$)) and lower
 300 ($\sim 10.181 \text{ MPa}\cdot\text{mm}^{0.5}$ (293 psi-inch $^{0.5}$)) SIF as in table 2 indicate that the butt-fusion joint
 301 material is susceptible to SCG failure mechanism over a large range of applied SIF.
 302
- 303 • Fracture parameters, CMOD and COA obtained with crack growth from the parent and
 304 butt-fusion joint HDPE SENT tests, show the larger material fracture resistance exhibited
 305 by the parent HDPE material.
 306
- 307 • Although bimodal HDPE material is much improved over older PE resins, the markedly
 308 lower SCG resistance of butt-fusion joint material persists. This presents an issue when
 309 considering the design and approval of these materials for nuclear power plant
 310 applications.

311 **REFERENCES**

312 [1] Barbee, T. W., Jr., L. Seaman, R. Crewdson, and D. R. Curran, 1972. "Dynamic Fracture
 313 Criteria for Ductile and Brittle Metals," *J. of Materials* 7,393-401. (UNCLASSIFIED).

314

315 [2] Carroll, M. M., and A. C. 1972. Holt, "Static and Dynamic Pore-Collapse Relations for
316 Ductile Porous Materials," *J. Appl. Phys.* 43(4), 1626-163. (UNCLASSIFIED)

317 [3] Cohen, L. J. and Berkowitz, H.M 1971 "Time-Dependent Fracture Criteria for 6061-T6
318 Aluminum Under Stress-Wave Loading in Uniaxial Strain," *Int. J. Fract. Mech.* 7(2), 183-196.
319 (UNCLASSIFIED)

320

321 [4] Fineberg, J.; Gross, S. P.; Marder, M. and Swinney, H. L 1992. Instability in the Propagation
322 of Fast Cracks. *Phys. Rev. B* **1992**, 45, 5146–5154.

323

324 [5] Kalyanam,S, Krishnaswamy,P, Hioe Y,2016. Slow Crack Growth Fracture Resistance
325 Parameter Evaluation For Parent And Joint Hdpe Materials *Engineering Mechanics Corp. Of*
326 *Columbus (Emc2), Columbus, Oh, & P. Raynaud, E. Focht, United States Nuclear Regulatory*
327 *Commission (Usnrc), Rockville, Md*

328

329 [6] Kalyanam. S, Krishnaswamy. P , Hioe.Y , Shim. D, Focht. E (2015). "A Fracture Mechanics
330 Approach to Service Life Prediction of HDPE Fusion Joints in Nuclear Applications," *Society of*
331 *Plastics Engineers, ANTEC2015*, Orlando, Florida, March 23-25

332

333 [7] Krishnaswamy P., Focht E., Shim D and Zhang T 2008. "Use of polyethylene (PE) pipe in
334 safety-related, Class 3, service-water piping", *Proc. of ASME 16th ICONE*, May 11-15, , Orlando,
335 Florida, USA.

336

337 [8] Krishnaswamy P., Kalyanam S. Hioe Y. and Shim D 2012. "A Methodology to Predict
338 Critical Flaw Size in HDPE Piping and Joints in Nuclear Safety-Related Applications," *Plastic*
339 *Pipes XVI*, Barcelona, Spain

340

341 [9] Morrissey, J.W. and Rice, J.R., 1996. 3D Elastodynamics of cracking through heterogeneous
342 solids: crack front waves and growth of $\bar{\nu}$ uctuations (Abstract). Transactions of the American
343 Geophysical Union 77(46, Fall Meeting Supplement), F485.

344 [10] Morrissey, J.W and Rice, J.R., 1998. Crack front waves. *Journal of the Mechanics and*
345 *Physics of Solids* 46, 467±487.

346

347 [11] Morrissey, J.W and Rice, J.R. 1999. Perturbative simulations of crack front waves. *Journal*
348 *of the Mechanics and Physics of Solids* (in press).

349

350 [12] Nakamura, T., Shih, C.F and Freund, L.B., 1985. Elastic-plastic analysis of a dynamically
351 loaded circumferentially notched round bar. *Engineering Fracture Mechanics* 22, 437±452.

352

353 [13] Nakamura, T., Shih, C.F and Freund, L.B., 1988. three-dimensional transient analysis of a
354 dynamically loaded three-point-bend duc-tile fracture specimen. In: *Nonlinear Fracture*
355 *Mechanics, Special Technical Publication 995*. American Society for Testing and Materials,
356 Philadelphia, pp. 217±241.

357
358 [14] Ravi-Chandar, K. and Knauss, W.G., 1984a. An experimental investigation into dynamic
359 fracture: I. crack initiation and arrest. *International Journal of Fracture* 25, 247±262.
360
361 [15] Ravi-Chandar, K. and Knauss, W.G., 1984b. An experimental investigation into dynamic
362 fracture: II microstructural aspects. *International Journal of Fracture* 26, 65±80.
363
364 [16] Ravi-Chandar, K., Knauss, W.G., 1984c. An experimental investigation into dynamic
365 fracture: III. On steady state crack propa- gation and crack branching. *International Journal of*
366 *Fracture* 26, 141±154.
367
368 [17] Rice, J.R., Ben-Zion, Y and Kim, K.-S., 1994. Three-dimensional perturbation solution for a
369 dynamic planar crack moving unsteadily in a model elastic solid. *Journal of the Mechanics and*
370 *Physics of Solids* 42, 813±843.
371
372 [18] Rittel, D and Levin, R. 1998. Mode-mixity and dynamic failure mode transitions in
373 polycarbonate. *International Journal of Mechanics of Materials* 30, 197±216.
374
375 [19] Rittel, D and Rosakis A. J 2005. Dynamic Fracture of Beryllium-Bearing Bulk Metallic
376 Glass Systems: A Cross-Technique Comparison. *Eng. Frac. Mech*, 72, 1905–1919.
377
378 [20] Shim D., Krishnaswamy P. and Focht E 2009. “Comparison of Parent and Butt Fusion
379 Material Properties of High Density Polyethylene”, *Proc. of ASME 2009 PVP Conf.*, July 26-30,
380 , Prague, Czech Republic
381
382 [21] Xu, W.; Yao, X. F.; Yeh, H. Y and Gin, G. C 2005. Fracture Investigation of PMMA
383 Specimen Using Coherent Gradient Sensing (CGS) Technology. *Polym. Test.* 24, 900–908.
384
385 [22] Arrhenius S 1889, On the reaction velocity of the inversion of cane sugar by acids,”
386 *Zeitschrift für Physikalische Chemie* 4, 226.
387
388
389
390
391
392
393
394
395
396
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