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

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Authors' contributions

This work was carried out in collaboration between authors PSE and CCI. Author PSE designed the study, performed the statistical analysis, wrote the protocol and wrote the first draft of the manuscript. Authors PSE and CCI managed the analyses of the study. Author PSE managed the literature searches. Both authors read and approved the final manuscript.

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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.

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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 [1]. Failure can be described as any change of properties which make the material or component functionally, structurally or aesthetically unacceptable [2]. 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 [3]. 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” [1].

Failure is defined as the separation of a body into two or more parts by tension or compression [4,1]. Dynamic failure of engineering materials (ductile metals) in tension takes place by the nucleation, growth and coalescence of voids [4]. For brittle materials, it takes place by the nucleation, growth, and coalescence of cracks [4]. As the rate at which materials are deformed increases, the following effects play an increasingly important role: (a) Mass inertia which leads to the propagation of elastic, plastic, and shock waves. (b) Thermal inertia, which is the thermal diffusion distance decreases as the time for deformation decreases, leading to pronounced temperature in homogeneities within the material. (c) Thermal activation and viscosity, which is the response for dislocations (the primary carriers of plastic propagation velocities phonon and electron viscosity may determine the response [1,3]. these three effects determine the elastic, plastic and failure response of materials. The infinite complexity of the morphological characteristics of failure can be rationalized by the interplay of the above-named effects (mass inertia, thermal inertia, thermal activation and viscosity) with microstructural characteristics of materials [4]. Dynamic failures of High-Density Polyethylene (HDPE) engineering materials can be classified into three groups [5]:

(a) Tensile failure: Which is the state of stress and the dynamics of generation, propagation, and interconnection of flaws dictates the morphology. Under uni-axial strain conditions, this failure is called "spalling" [1].

(b) Compressive failure: Under compressive traction localized regions of tension can be generated in the microstructure, which gives rise to failure. Although metals (gold, silver, etc.) are mostly immune to this type of failure, less ductile metals (eg., tungsten, steel), ceramics, composites are subjected to this type of failure [5].

(c) Shear failure of shear localization: With usually micro-structural or thermal origins, often leads to failures. It should be emphasized that the adiabatic shear band is the precursor event and that it provides a path for crack propagation (fragilized or softened material) which is a tensile stress phenomenon. Failure of materials may have huge costs. Causes included improper materials selection or processing, the improper design of components, and improper use.

Failure in polymer components can occur at relatively low stress levels (far below the tensile strength in many cases) due to long-term stress (creep-rupture), cyclic stresses (fatigue failure) or liquid agents (environmental stress cracking) [2]. When a polymer is stressed in the air to just below its yield point, stress cracking can occur after some time. However, when simultaneously exposed to both stress and a chemical medium, this will result in a sharp reduction of the time to failure [6]. This type of failure has been named environmental stress cracking (ESC). ESC has been a subject of extensive investigations for almost 50 years [2]. It has deserved much attention because approximately 15 to 20 % of all failures of plastic components in service are due to ESC [7].

At this point, the industry was confronted with numerous reports of polyethene failure. Polyethene was reported to be unsatisfying for cable usage, and it was found to crack violently on contact with methanol at room temperature. Therefore, the problem of ESC is very important for many applications including packaging industry (bottles, containers, foils, films, etc.), electric industry and electronics (wire and cable insulation), medicine (labware, caps, implant components, etc.), automobile industry (tanks, pipes, coatings, etc.) and many more [4].

Therefore, Dynamic Failure of High-Density Polyethylene (HDPE) engineering material is a form of (silent) failure where the material separates in pieces due to stress at temperatures below the melting point [2]. The failure is termed ductile or brittle depending on whether the elongation is large or small. Steps in failure (response to stress) are Track formation and Track propagation [8]. A common characteristic of these failure phenomena is the rapid loss of stress carrying capability in time scales as depicted in Table 1 [9,10].

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

1.1 Ductile Fracture

In High-Density Polyethylene (HDPE) material, the material fracture slowly and deform plastically before seeing a sign of a crack that is because those parts are built in a ductile material as in Fig. 1 [11]. Stages of ductile fracture are; Initial necking, small cavity formation (microvoids), Void growth (ellipsoid) by coalescence into a crack, Fast crack propagation around the neck. Shear strain at 45º final shear fracture (cup and cone), The interior surface is fibrous, irregular, which signify plastic deformation [10].

Ductile fracture with characteristic distortion and shear lip, steel is a ductile material. It extends and deforms before failing as in Fig. 2b and 2c below [10]. That is represented in the stress-strain curve as in Fig. 3. These Fig. 2a, 2b, 2c depicts the difference between brittle, ductile and semi-ductile materials respectively:

1.2 Brittle Fracture

In this failure, there is no appreciable deformation and crack propagation is very fast as in Fig. 2a. In most brittle materials, crack propagation (by bond breaking) is along specific crystallographic planes (*cleavage* planes) [9]. This type of fracture is transgranular (through grains) producing grainy texture (or faceted texture) when cleavage direction changes from grain to grain. In some materials, the fracture is intergranular [10].

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

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

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

Fig. 1. Depicts the failure of ductile materials

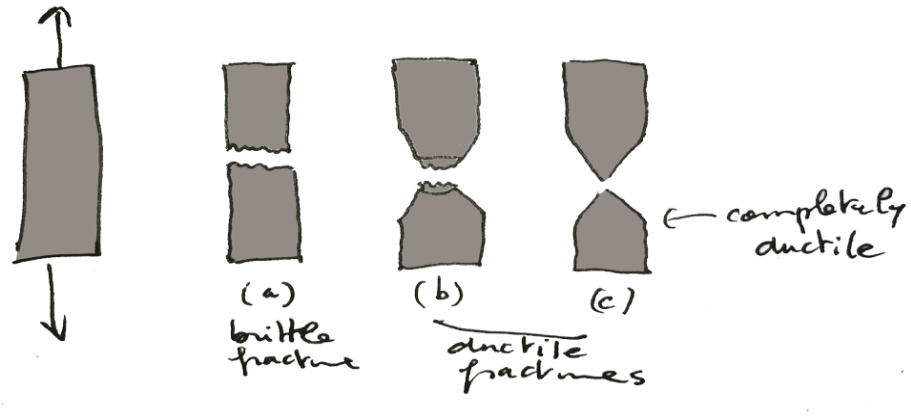

Fig. 2. Depicts the difference between brittle, ductile and semi-ductile materials

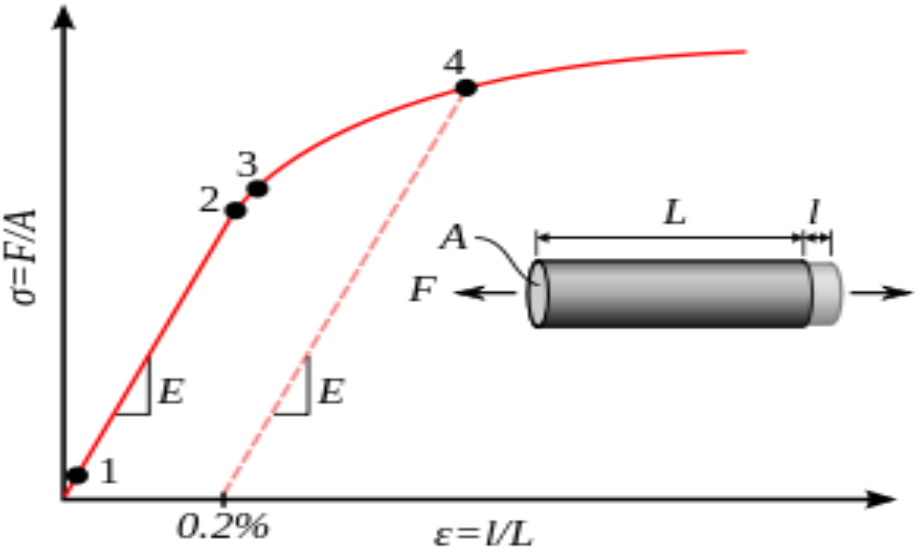

Fig. 3. Typical yield behaviour of engineering materials

Elastic limit (yield strength): Beyond the elastic limit, permanent deformation will occur. The elastic limit is, therefore, the lowest stress point at which permanent deformation can be measured. This requires a manual load-unload procedure, and the accuracy is critically dependent on the equipment used and operator skill. For elastomers, such as rubber, the elastic limit is much larger than the proportionality limit. Also, precise strain measurements have shown that plastic strain begins at low stresses [10].

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

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

Upper and lower yield points: Some materials, reach an upper yield point before dropping rapidly to a lower yield point. The material response is linear up until the upper yield point, but the lower yield point is used in structural engineering as a conservative value. If the material is only stressed to the upper yield point, and beyond, Lüders bands can develop [10]. When a propagating crack tip passes a material point, the material point instantaneously separates into at least two parts. In the ordinary numerical models with nodal release techniques,

this sudden unloading process often produces spurious oscillations. The crack propagation velocity along a bimaterial interface can become extremely fast and can exceed the shear wave velocity of the compliant material [11].

1.3 Crack Initiation and Propagation

Stages in fatigue failure are categorized as follows: (I). Crack initiation at high-stress points (stress raisers). (II). Propagation (incremental in each cycle). (III). Final failure by fracture [3].

Stage I – propagation

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

Stage II – propagation

- Crack propagates by repetitive plastic blunting and sharpening of the crack tip [6].
- Crack Propagation Rate (not covered)

2. HOW STRESS CRACKING OCCURS IN ENGINEERING MATERIAL

The failure of a container to resist ESC can be the result of stored stresses acquired in the moulding or extrusion operation [12]. Stress cracking agents, such as the liquids mentioned above, migrate into minute cracks in the crystalline areas of the polyethene molecules forming the surface of the container. These microscopic cracks are a result of a breakdown of the polymer chains in the case of acids and solvents, and a “wetting out” of the surface in the case of detergents, largely due to their surfactant components [2].

In either case, the surface tension between the crystalline layers is reduced. What happens is that once a microscopic surface imperfection propagates or "zippers open" to a full-fledged break in the bottle. ESC failures are accelerated by high temperatures and additional external stresses such as top-loaded storage [13].

3. SIGNIFICANCE OF ENVIRONMENTAL STRESS CRACKING RESISTANCE

3.1 The Stress Factor

As the name suggests, stress cracking requires the polymer to have exposure to intrinsic residual stress or an externally applied stress. If the plastic moulding is completely free of stress, then no stress cracking will occur [2]. Even polymers exposed to liquids or vapours that have a swelling or wetting effect will not undergo ESC unless there is an externally applied or moulded-in stress present. External stress may be the result of component assembly (composite formulation), improper packing or storage, incorrect use, etc [14].

The definition of cracking due to stress is defined in many standards. It is stated as the internal or an external crack in the plastic caused by stresses less than its short-term mechanical potency [15]. This type of cracking usually consists of brittle cracking with no ductile or little drawing of plastic material from the adjoining surface failure. Slow growth in cracks is another term used to explain stress cracking [16].

Environmental stress cracking (ESC) in plastics means the failure at about room temperature due to continuously acting external and/or internal stresses in the presence of surface-active substances (known as stress cracking agents) such as alcohols, soaps, dyes, agents containing moisture [17]. Although ESC results from the interaction of the polymer with certain chemicals, it is not a chemical reaction between the polymer and the active environment [18]. The stress cracking agents do not cause any chemical degradation of the polymer but they accelerate the process of macroscopic brittle-crack formation.

4. REVIEW OF RELATED LITERATURE

Several authors [19,15], used a modified ENF specimen to determine the mode II dominated dynamic delamination fracture toughness of fibre composites at high crack propagation speeds. A strip of adhesive film with higher toughness was placed at the tip of interlaminar crack created during laminate lay-up [6]. The objective was to delay the onset of crack extension and produce crack propagation at high speeds (700 m/s). Sixteen pure aluminium conductive lines were put on the specimen edge side using the vapour deposition technique, to carry out crack speed measurements [13]. The authors concluded that the mode II dynamic energy release rate of unidirectional S2/8553 glass/epoxy composite seems to be insensitive to crack speed within the range of 350 and 700 m/s. The authors also simulated mixed mode crack propagation by moving the pre-crack from the mid-plane to 1/3 of

the ENF specimen thickness of unidirectional AS4/3501-6 carbon/epoxy laminates [13,16].

The majority of the experimental studies consider unidirectional laminates. [2,12,19,15], performed an experimental investigation of dynamic crack initiation and growth in unidirectional fibre-reinforced polymeric-matrix thick composite plates. Edge-notched plates were impacted in a one-point bend configuration using a drop-weight tower. Using an optical method the authors carried out a real-time visualization of dynamic fracture initiation and growth for crack speeds up to 900 m/s. They verified that the elastic constants of the used material are rate sensitive and the measured fracture toughness values are close to those types of epoxies. This was considered consistent because in unidirectional lay-ups crack initiation and growth occurs in the matrix.

Several authors [6,17], have suggested that the dynamic fracture behaviour of materials depends on the balance between the energy released by the structure over a unit area of crack propagation (*G*) and the material resistance (*R*), which can be viewed as the energy dissipated in creating the fracture surface. When unstable crack growth occurs, the difference *G*-*R* is converted into kinetic energy. If *G* increases with crack growth the crack speed also increases because more energy is available. The crack arrest will occur when *G* becomes lower than *R* and, consequently, no kinetic energy is available for crack growth. Thus, it can be affirmed that fracture stability depends on the variations of the strain energy release rate and the resistance of the material during crack growth.

5. STUDY BACKGROUND

5.1 Fatigue Failure Propagation

Many materials will fail at lower stress when subjected to cyclic or repetitive loads than when under static loads [5]. And it has been found experimentally that when a material is subjected to repeated stress; it fails at stresses below the yield point stresses. This type of failure as regards engineering materials is known as fatigue [4]. Fatigue failure is caused through progressive crack formation which is usually fine and of microscopic size, which occurs even without any prior indication. For thermoplastics pipe materials, fatigue is only relevant where a large number of stress cycles are anticipated [2]. The important factors to consider are the magnitude of the stress fluctuation and the loading frequency. Where large stress fluctuations are predicted, fatigue design may be required where the total number of cycles in the operational lifetime of the pipe exceeds 100,000. For smaller stress cycles, a larger number of cycles can be tolerated [12].

5.2 Modeling of Slow Crack Growth (SCG) Resistance of HDPE Parent Material

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

The Brown model employs a power-law function of SIF (*KI*), an exponential function of temperature (akin to an Arrhenius equation [2,15]), and functions to capture geometric dimensions/constraint factors. The model, as shown in equation (1) for life prediction bases its approach on the PENT failure time and has been used extensively by the plastic pipe industry and has recently been considered by the nuclear industry [2,18,15].

$$tf = [\text{tPENT}] \left(\frac{0{,}468}{KI}\right)^{n} Exp\left[\frac{Q}{R}\left(\frac{1}{T} - \frac{1}{353}\right)\right] \qquad (1)$$

where *tf* is time to failure (service life)

tPENT = PENT failure time

KI = stress intensity factor (=0.468 MPa-m$^{0.5}$ for PENT specimen)
n = stress intensity exponent (~2.5 to 4.5)
Q is resin activation energy (~85 to 110 kJ/mol)
R is universal gas constant 0.008314 kJ/mol/°K, and *T* is absolute temperature in °K (=353 °K for PENT specimen)

The Brown model was calibrated using the slope *n*=3.25, PENT value *tPENT*=60 hours, and the temperature shift function, *Q*=100 from test data for the HDPE PE4710 bimodal butt-fusion joint materials [2,7,15,20].

5.3 Experimental Investigation on the SCG Resistance of HDPE Parent Material

The service life prediction models employ the failure time from a PENT, SENT [19] test to

obtain the performance life of HDPE material with a detected flaw depth, and operating temperature, and pressure [2,17,7]. Recent work indicate that the PENT and SENT specimens to have a SIF and constraint/transverse *T*-stress (coefficient $\beta = [T(\pi\alpha)^{0.5}]$ /KI) that is close to that of a surface crack in an HDPE pipe and have been the focus of the investigations [2,17,21,11].

To determine the variability of failure times and validate service model predictions; tests were conducted at PENT *KI* = 14.942 MPa-mm$^{0.5}$ (430 psi-inch$^{0.5}$), SENT tests at *KI* = 18.486 MPa-mm$^{0.5}$ (532 psi-inch$^{0.5}$), and at *KI* = 10.181 MPa-mm$^{0.5}$ (293 psi-inch$^{0.5}$) [2,7,20].

Comparison of parent material SCG failure time (7893 hours when tested at 95°C and *KI* =18.486 MPa-mm$^{0.5}$ (532 psi-inch$^{0.5}$)) with the butt-fusion joint material SCG failure time (less than 100 hours when tested at 95°C and *KI* = 18.486 MPa-mm$^{0.5}$ (532 psi-inch$^{0.5}$) were discussed in [2,12,11]. Table 1 shows the list of butt-fusion joint SENT and PENT specimens tested from the 10k and 2k PE4710 HDPE resins and their corresponding failure times were evaluated as in equation 1 [2,11,19].

The four stages of SCG in the 10k-P3 parent HDPE specimen that was correlated to the observations depicts initial crazing, accumulation of damage, the start of SCG, and final failure [2,17,19,20]. The experimental data obtained during the crack growth over several hours of creep test time using equation 1 to model the crack-mouth-opening displacement (CMOD) and crack-opening angle (COA) as a function of time. The crack length at various times during the creep test was obtained by considering the crazed region and fibril breakage in the vicinity of the current crack front as seen on the surface of the HDPE SENT specimen [2,17,19] and the crack length measurements in these tests were chosen to be until the end of the crazing zone.

6. RESULTS AND DISCUSSION

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

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

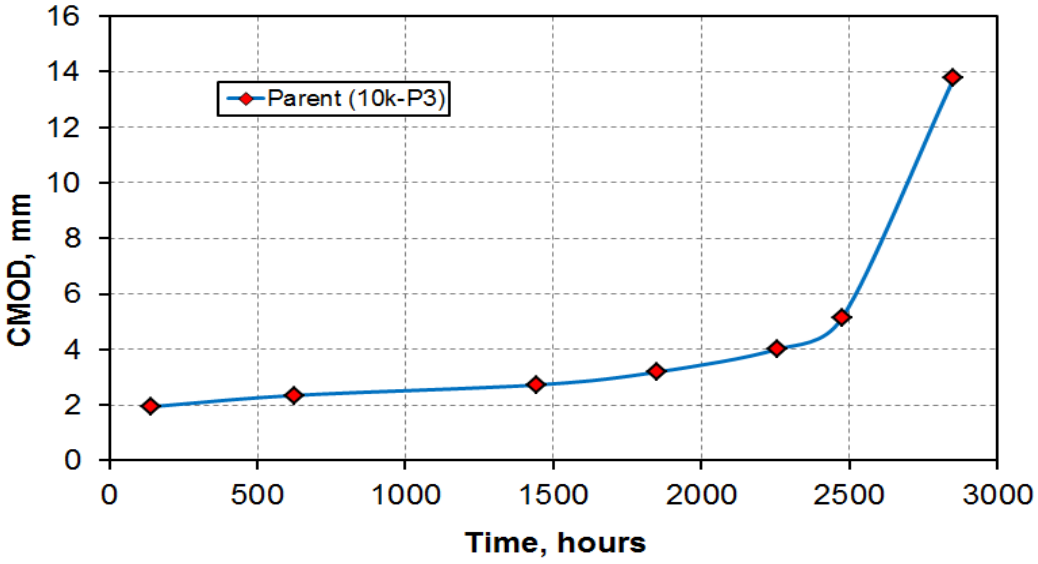

Fig. 4. CMOD versus time during SCG in parent HDPE SENT specimen, 10k-P3

Table 2. SCG tests on Butt-fusion Joint (B) Materials of PE4710 Bimodal HDPE 10,000 hours PENT (10k) and 2,000 hours PENT (2k) Resins [2,11]

Specimen Name	Test Frame	SIF, KI MPa-mm$^{0.5}$ (psi-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

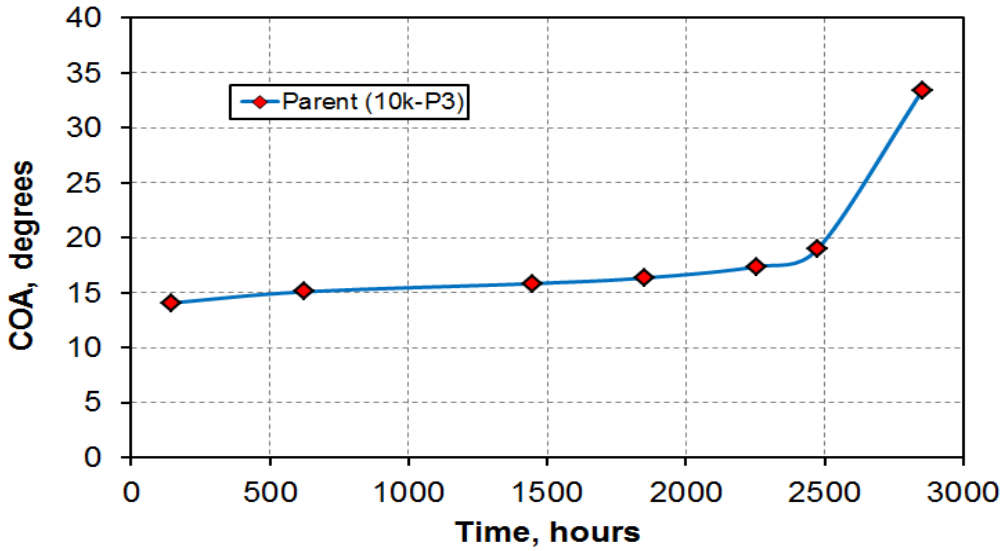

Fig. 5. COA versus time during SCG in parent HDPE SENT specimen, 10k-P3

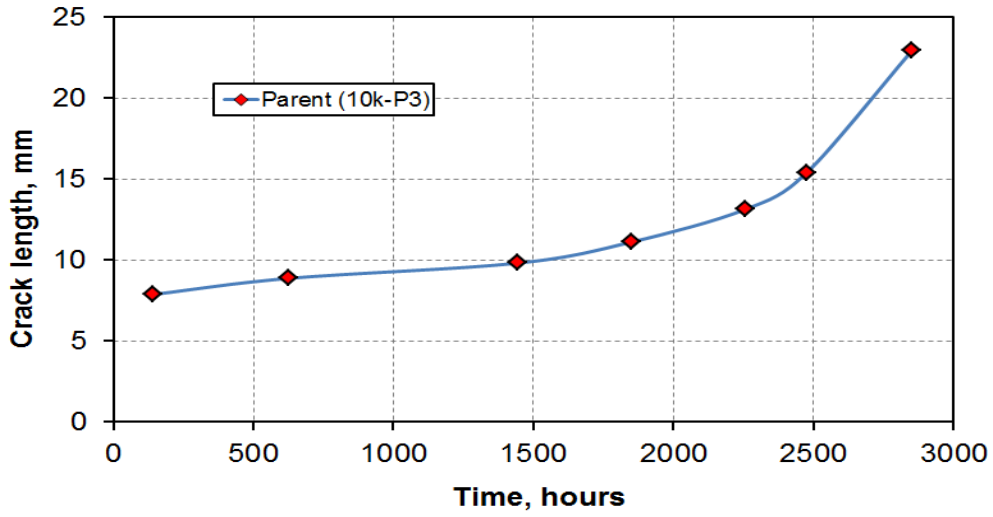

Fig. 6. Crack length versus time during SCG in parent HDPE SENT specimen, 10k-P3

7. CONCLUSIONS

The following conclusions were drawn from this work:

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

COMPETING INTERESTS

Authors have declared that no competing interests exist.

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