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Crashworthiness characteristics of a carbon fiber reinforced dual-phase epoxy-polyurea hybrid matrix composite

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Abstract:

The crashworthiness characteristics of rectangular tubes made from a carbon-fiber reinforced hybrid-polymeric matrix (CHMC) composite were investigated using quasi-static and impact crush tests. The hybrid matrix formulation of the CHMC was created by combining an epoxy-based thermosetting polymer with a lightly crosslinked polyurea elastomer at various cure-time intervals and volumetric ratios. The load-displacement responses of both CHMC and carbon-fiber reinforced epoxy (CF/epoxy) specimens were obtained under various crushing speeds; and crashworthiness parameters, such as the average crushing force and specific energy absorption (SEA), were calculated using subsequent load-displacement relationships. The CHMC maintained a high level of structural integrity and post-crush performance, relative to traditional CF/epoxy. The influence of the curing time and volumetric ratios of the polyurea/ epoxy dual-hybridized matrix system on the crashworthiness parameters was also investigated. The results reveal that the load carrying capacity and total energy absorption tend to increase with greater polyurea thickness and lower elapsed reaction curing time of the epoxy although this is typically a function of the loading rate. Finally, the mechanism by which the CHMC provides increased damage tolerance was also investigated using scanning electron microscopy (SEM).

Keywords: *A. Polymer-matrix composites (PMCs); A. Hybrid; B. Impact behavior; B. Damage tolerance; Polyurea*

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1. Introduction

The design of energy absorbing structural materials has become an increasingly investigated topic in the manufacturing of crashworthy structures. A crashworthy structure is one that is designed to absorb significant impact energy such that the resultant force and energy that are transmitted to, for example the passengers in a vehicle, are minimized. Because of their capability to deform plastically, metallic materials have been commonly used in crashworthy applications for several decades [1]. However, because of their many merits, including high specific stiffness/strength and superior durability, fiber reinforced polymers (FRPs) has become a popular choice in vehicular applications [2-4]. Unlike metals, most FRP materials do not exhibit significant plastic deformation; instead, the manifestations of energy absorption consist of internal damping, which results in energy dissipation capacity, and damage tolerance, which is the ability of materials to develop hierarchical damage without losing significant load carrying capacity. Studies have also revealed that the primary energy dissipation mechanism in composite materials is due to their synergistic sequence of failure mechanisms [5;6].

To date, a number of research studies have focused on initiating a better understanding of the energy absorption capacity of composite materials and structures. Research on the energy absorption of composite structures traces back to the experimental test results of Thornton [7], which demonstrated higher specific energy absorption (SEA) in axially crushed composite tubes over their metallic counterparts; interestingly, most of the tested tubes failed in brittle fracture mode. In 1983, a comparative study was performed by Farley [8], investigating the energy absorption characteristics of composite tubes using graphite/epoxy, Kevlar/epoxy, and glass/epoxy pre-impregnated woven fabric. A total of 128 specimens with chamfered, or notched,

edges were tested under both static and dynamic load conditions. The results revealed a clear dependency of material energy absorption on the fiber orientation and the type of reinforcing fiber used. A review performed by Jacob et al. [9] summarized the influence of several factors, such as fiber type and fiber orientation, matrix material, specimen geometry, processing conditions, and loading rate on the energy absorption capacity of specimens. A large spectrum of experimental test results of variously configured specimens were subsequently analyzed. Certain criteria specifically targeted for automobile design, including deceleration during crushing, were included in the comprehensive study. More recently, research efforts have examined the energy absorption and crush performance of many types of emerging composite materials such as nanocomposites [1;10;11] and composite materials reinforced with natural fibers such as silk [12-14]. Studies on nanocomposites in particular reveal that many types of nano-fillers, e.g., carbon nanotubes (CNTs) and nano-graphane, are effective in increasing the toughness and energy absorption capability of composites, however, the processing cost of such materials is generally high. Polymeric composites reinforced by natural fibers, e.g., silk and sisal fibers, exhibited satisfactory impact resistance and energy absorption. However, one obstacle that has been identified in natural composite applications is environmental durability [9].

In many crashworthy applications, such as those applicable to the automotive industry, energy absorption capability may not be the only concern since cost and durability also play a substantial role. In an attempt to develop more cost efficient and durable composite materials with increased damage tolerance, a polymeric matrix composite (PMC) that was recently developed by some of the authors [15-19] has been

experimentally tested herein, in order to determine the energy absorption and crashworthiness characteristics of the material. This PMC is referred to as Carbon-fiber reinforced Hybrid Matrix Composite, or CHMC [20], consisting of high strength carbon fibers and a dual-phase hybridized matrix composed of Bisphenol A (BPA) epoxy and an amino-based cross-linked polymer to formulate a polyurea compound that ultimately helps provide high damping, fracture toughness, and superior strength sustainability. Earlier studies conducted by Zhou and Attard [12] and Zhou et al. [16;17] signify tremendous potential of the CHMC as a retrofitting material to sustain the strength of otherwise brittle substrates. Since thin-wall rectangular tubes are popularly used for automobile structures, such as the door beams and frame rails [21], the performance of CHMC tubular sections is experimentally investigated under low-speed impact loads (2 - 5 m/s). Rectangular tube specimens that are manufactured using CHMC designed with various formulations were tested under both quasi-static and dynamic load conditions. The results, including the load-deflection relationships and crashworthiness parameters, are presented. The influence of the two material processing parameters - the polyurea thickness, h_p , and elapsed reaction curing times of the epoxy matrix phase, t_c , - with respect to the performances of the CHMC tubes under crush loads are assessed. The results are also compared to those obtained using carbon fiber reinforced epoxy specimens. Lastly, a fractography study was used to analyze the mechanism by which the CHMC stabilizes damage and is able to sustain substantial post-peak load carrying capacity, whereas traditional carbon-fiber reinforced epoxy was shown to be markedly deficient in these quantitative measures.

2. Crashworthiness characteristics of fibrous composite structures

Crashworthiness is defined as the ability of a vehicle structure to mitigate load transfer to occupants and to alleviate serious injury or death in the event of an accident [9]. The load transfer to occupants is minimized by dissipating the kinetic energy of a moving vehicle through energy absorbing structural materials or devices. The ability of a vehicular structure to dissipate energy as well as maintain structural integrity in an accident is affected by both the geometry of the crushed structure and the mechanical properties of the constituent materials, where the energy dissipation capability of the role of the latter is vital in the overall performance of vehicle structures. The energy is primarily dissipated by plastic deformation in metallic structures, whereas in composite structures, it may be dissipated by various sources, including the viscoelastic/viscoplastic nature of the load-bearing fibers and matrix constituents, the interaction between material phases, and the various damage mechanisms [22]. A number of studies on the impact performance and energy absorption of composite structures has identified differences in failure modes and crash patterns between metallic and composite structures. Some of the primary crushing modes in continuous fiber reinforced composite structures include 1) transverse shearing or fragmentation, 2) brittle fracturing of the fibers or matrix, 3) lamina bending or splaying, and 4) local buckling or progressive folding [23]. Closer examination of each mode reveals that: the fragmentation mode occurs primarily in brittle fiber reinforced tubes and is typically characterized by a wedge-shaped laminate cross section with short interlaminar and longitudinal cracks; the splaying mode is characterized by very long interlaminar or intralaminar parallel-to-fiber direction cracks; the brittle fracturing mode is a combination of the fragmentation and splaying modes; and lastly, progressive folding is characterized by local buckling of the fiber components or

matrix brought on by plastic yielding which drives the energy dissipation mechanism of the composite.

Figure 1 presents the force-displacement relationship obtained from a crush test, where three stages can be distinguished from the response curve: the pre-crush stage which features the initiation of failure in the material; the post-crush stage which occurs when material damages spreads across the specimen; and finally the compaction stage, during which residual debris is compacted [24]. Using the force-displacement responses of test specimens, the values of certain descriptive parameters are calculated in order to quantitatively evaluate the crashworthiness characteristics of the composite components. These include:

- Peak force (F_{max}) which is defined as the maximum load during the pre-crushing stage, see Figure 1;
- Average crushing force (F_{ave}), or the sustained load, which is defined as the average crushing force during the post-crush stage, obtained by equation (1) as

$$F_{ave} = \frac{1}{\Delta} \int_0^{\Delta} F d\delta, \text{ when } F > F_{max} \quad (1)$$

where F is the force; δ is the displacement; and Δ is defined as the post-crush displacement, see Figure 1;

- Energy absorption (EA) which is calculated as the area under the force-displacement curve and is expressed as a function of the displacement:

$$EA = \int_0^{\delta} F d\delta \quad (2)$$

- Specific energy absorption (SEA) which is the energy absorbed per unit mass of the structure being crushed;

- Crush force efficiency (CFE), which is the ratio of the average crushing force, F_{ave} , and the peak force, F_{max} .

3. The CHMC material and Test method

3.1. The carbon-fiber reinforced hybrid polyurea-epoxy matrix composite

While the strength and stiffness of continuous fibrous composites are mostly governed by the fiber properties, the matrix phase provides stress- transferring paths for the reinforcing fibers; thus, the matrix phase, at least to some extent, affects the toughness of composites [25] and is a conduit for energy dissipation. Most thermosetting polymers, such as epoxies, fracture at relatively low ultimate strains; as a result, damage that may initiate in the reinforcing fibers or at the fiber-matrix interface quickly bridges together with other damage modes initiated in other fibers or in fiber-matrix interfaces via matrix cracking. The result is catastrophic failure of the material. An approach to increase damage tolerance of polymeric composites is to induce "damage barriers" so that (local) microscopic damage events are isolated, rendering any single crack harmless [17]. To develop this concept, a Carbon-fiber reinforced Hybrid Matrix Composite, or CHMC, has been recently developed to improve the damage tolerance and energy absorption capacity of otherwise brittle carbon-fiber reinforced polymers. The polyurea-epoxy dual hybridized matrix system of CHMC is created by saturating the reinforcing carbon fiber in a Bisphenol A (BPA) diglycidal ether epoxy matrix phase I, where the condensation crosslinking reaction between epoxy resin molecules and the hardener takes place to form a highly crosslinked reaction product. Following an elapsed amount of reaction curing time (denoted here as t_c) of the phase I agents, a mixture of methylene diphenyl diisocyanate (MDI, part A) and a polyamine hardener (part B) at a volume ratio of part A

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4 and part B of 1.05 was applied to the surface of the matrix phase I to form a lightly
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6 crosslinked polyurea elastomer layer of thickness h_p . The epoxy prepolymer and
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8 polyurea react via the reaction of the polyurea end group (NCO) and the hydroxide group
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10 of the epoxy, or via the reaction of amine with the epoxide group [26-28], forming a well-
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12 bonded interfacial layer between the two polymeric phases that provides the necessary
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14 energy dissipation mechanism by which ultimately single damage events that may have
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16 occurred do not grow unstably, and by which damping also increases. The scanning
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18 electron microscopy (SEM) images presented in Figure 2 show the multilayered cross-ply
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20 texture of CHMC, where reinforcing carbon fibers are saturated in an epoxy matrix phase,
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22 followed by a surface coating application of the elastomeric polyurea matrix phase II
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24 onto the laminate (on either one side of it or on both sides). The interface between the
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26 epoxy and polyurea appears to provide an excellent bond, see Figure 2d, where the width
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28 of this interface is a function of the reaction time between the polyurea and the epoxy
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30 resin.
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38 The specimens that were used to investigate the crashworthiness of CHMC and
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40 the carbon-fiber reinforced epoxy consisted of rectangular composite tubes fabricated
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42 using 2×4×9 inch polystyrene foam cores using a hand lay-up process, see Figure 2e.
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44 The specimen cross-section size (2×4 inches) was selected based on common cross-
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46 section dimensions used for race car frame rails [21]. The reinforcing fibers consist of a
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48 Torayca[®] unidirectional carbon fiber fabric having a nominal single-ply thickness of 0.33
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50 mm. The test specimens were constructed using single-ply laminas reinforced
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52 unidirectionally (longitudinally), see Figure 2e; the epoxy resin and hardener were
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54 produced by BASF, The Chemical Company. The properties of the constituent carbon
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4 fiber and epoxy are reported in the manufactures' data sheets, where the elastic modulus
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6 of cured epoxy resin is reported as 3,034 MPa [29;30]. The elastic moduli of the two
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8 polymeric matrix phases were tested using nanoindenter, where the elastic modulus of
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10 the polyurea phase was tested as 130 MPa. Additionally, the cured polyurea polymer
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12 used in this study is reported as 110-500% elongation prior to fracture [31].
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15 16 17 **3.2 Test setup and instrumentation** 18

19 Both quasi-static and impact crush tests were used to evaluate the energy
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21 dissipation/absorption capacity of the CHMC tubular specimens. Two quasi-static axial
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23 compressive tests were performed using an MTS 810 universal test machine having a 100
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25 kN load capacity. The specimens were loaded in a displacement controlled mode at a
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27 cross-head speed of 7.6 mm/min (or 0.005 in/s), and the load and displacement data was
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29 recorded throughout. The impact crush tests were performed using the Test Machine for
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31 Automotive Crashworthiness (TMAC, MTS Corp) at the Oak Ridge National Laboratory
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33 (ORNL). TMAC is a closed-loop controlled servo-hydraulic high speed testing system
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35 having a maximum load capacity of 490 kN, see Figure 3a. The maximum loading rate
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37 (stroke traveling speed) of the TMAC system is 8 m/s. The specimens were tested in a
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39 contained test chamber, where a Photron FASTCAM SA5 high-speed camera was staged
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41 to record the crushing process, see Figure 3b.
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48 Rectangular tube specimens were made from CHMC designs having various t_c
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50 and h_p values were tested using the TMAC. The selection of elapsed curing time - t_c - ,
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52 and polyurea thickness - h_p - was based on the expected influence on properties, such as
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54 damping, of the CHMC, which were identified in previous studies [17;19]. It was found
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56 from previous study [19] that the material damping and impact resistance generally
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increase with lower t_c values in CHMC systems at higher load rates, where the influence of h_p lessens; however, at lower load rates, these properties become more dependent on both h_p and t_c . Similar to how the strain rate may affect the mechanical response of a material, crushing speed (or loading rate) also affects the load response and energy absorption capability of a composite structural element. Therefore, three loading rate levels, namely 2 m/s, 3 m/s, and 5 m/s, were used for testing. The load - displacement curves were recorded by the load cell and LVDT sensors installed on the TMAC system. The LVDT is located inside the actuator, and it was calibrated at ± 300 mm travel range; a 500 kN lightweight piezoelectric load cell and a DC strain gauge load cell were installed in series located both at the bottom platform, see Figure 3b. Automated data acquisition was performed using National Instrument hardware and a in-house LabView program. The energy dissipation and load carrying capacity of the test specimens were studied following computation of the crashworthiness parameters and the load-displacement relations; the crushing process and failure modes were examined using the high-speed camera at a sampling rate of 20,000 frames per second (fps). The detailed geometrical and material information of the specimens and their test conditions, e.g. loading rate, are summarized in Table 1.

4. Results and discussion

4.1. Force-displacement relations and the failure modes

Table 1 summarizes the primary test results, i.e., peak force, average crushing force, and energy absorption etc., of both the quasi-static and impact crush tests; the force-displacement responses obtained from the crush tests are presented in Figure 4 and Figures 5-7 for the quasi-static and impact tests, respectively. The quasi-static tests were

performed on specimens made using carbon fiber reinforced epoxy (CF/epoxy) and CHMC having $h_p = 1.59$ mm and $t_c = 2.5$ hours. The force-displacement responses obtained for these two types of materials are compared in Figure 4. The CF/epoxy tubular specimen lost its load carrying capability soon after the peak load point, showing premature failure caused by matrix cracking in the parallel-to-fiber (or longitudinal) direction; on the other hand, the CHMC specimen sustained about 45% of its peak load during the post-crush stage, see the solid black line shown in Figure 4. The test was terminated when the MTS 810 testing machine reached its maximum stroke prior to reaching the compaction stage (see Figure 1 and 4), implying that the energy absorption (EA) capacity for the CHMC specimen under static loading may be underestimated.

Figures 5-7 present the combined load/energy-displacement relationships of CHMC specimens that were tested under three loading rate levels, 2 m/s, 3 m/s, and 5 m/s, see Table 1. In addition, one carbon fiber reinforced epoxy specimen was labeled as the control group and tested at a loading rate of 5 m/s; the additional test was used to compare the failure modes and energy absorbing capacities of the CHMC and the CF/epoxy tubes. During the pre-crush stage, i.e., prior to reaching the peak load, see Figure 2, CHMC specimens behaved almost linearly, see Figures 5-7. The pre-crush stage was featured by the formation of local buckles near the point of impacts, see Figure 8a (1). Following the peak load point, the load-displacement curves indicate that the CHMC specimens maintained a substantial level of post-peak load carrying capability (about 35-70% of F_{max}) over a relatively large post-crush displacement, indicating large energy dissipation and good ductility. Images captured by the high speed camera showed local buckling and progressive folding of the CHMC lamina during the post-crush stage,

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4 see Figure 8a (2)-(4). Most CHMC specimens maintained their structural integrity during
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6 the crush process; and only well-bridged short cracks/crazing were observed on the
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8 matrix near the point of impacts. The post-crush integrity and local buckling/progressive
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10 folding failure mechanism may be a result of the hyperelastic and plastic characteristics
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12 of the hybrid matrix phase, whereby the impact energy was dissipated mainly by the
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14 plastic deformation of matrix materials. More details regarding the damage/failure
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16 mechanisms of CHMC specimens under impact will be discussed in later sections. In
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18 addition, it is observed from Figures 5-7 that the load-displacement behavior of the
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20 CHMC specimens varies with the loading speed. In general, the post-crush average force
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22 (F_{ave}) increases as the loading speed increases; however, higher loading speed resulted in
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24 lower post-crush displacement values, see Figure 5-7.
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31 In contrast to the ductile crushing of CHMC specimens, the carbon fiber
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33 reinforced epoxy specimen exhibited a splaying failure mode, featured by the propagation
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35 of very long parallel-to-fiber matrix cracks over the entire height of the specimen during
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37 the post-crush stage, see Figures 8b (2)-(4). Since the primary energy dissipation
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39 mechanism of the splaying failure mode is manifested primarily as crack growth and
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41 since the laminas that had failed by splaying exhibited significant bending deformation
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43 without fiber fracture (Figures 8b (3) and (4)), then, as a result, it can be inferred that the
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45 splaying failure mode may have led to inefficient energy dissipation of the CF/epoxy
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47 specimen. This is consistent with the fact that there is significantly less energy
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49 absorption (EA) capability associated with the CF/epoxy specimen as compared to its
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51 CHMC counterpart, see Table 1. In addition, the premature failure caused by the
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parallel-to-fiber crack growth has led to minimal post-crush displacements, which also contributed to less energy dissipation upon failure.

4.2. The crashworthiness parameters

The primary crashworthiness characteristics of the tested rectangular tubes, namely the peak force (F_{max}), average crushing force (F_{ave}), total energy absorption (EA), and specific energy absorption (SEA), are plotted in Figure 9 as functions of the two material processing parameters: t_c and h_p . The test results are sorted into two groups with alternative t_c values, e.g., $t_c = 2.5$ hrs versus $t_c = 3.5$ hrs (see the square and triangular symbol marks in Figure 9); and the results were plotted against the thickness of the polyurea matrix phase (h_p) in order to study the influence of the two material processing parameters on material mechanical properties .

4.2.1. Peak force and average crushing force

The peak force (F_{max}) is defined as the maximum load experienced by the specimens during the pre-crush stage, see Figure 2. Figures 9a and 9b indicate that both the peak force and the average crushing force increase as the thickness of the polyurea matrix phase (h_p) increases or as the epoxy curing time (t_c) decreases. According to a study by Garattoni [32], the peak and post-crush load carrying capacities of specimens undergoing progressive folding are mainly attributed to the flexural stiffness of the composite lamina. A larger lamina thickness leads to a larger bending stiffness, which allows the load carrying capacity of the composite tubular specimen to increase, and under quasi-static loads, this result correlates well to an higher h_p ; however, as the load rate increases, the influence on the peak and post-crush load carrying capacities may begin to shift inversely proportional to t_c , i.e. t_c , lower , within certain curing time period.

Therefore, the sudden release of energy that would otherwise go towards extending (or forming new) crack surfaces is controlled primarily by the epoxy curing time and to some extent by the thickness of the hyperelastic polyurea phase in the matrix system, resulting in an increase of fracture toughness in the composite. As a result, the structural integrity of the specimens may be better maintained throughout the crushes, resulting in larger load carrying capacity.

The test results presented in Figure 9 were also grouped according to different t_c values, namely for $t_c = 2.5$ hrs and $t_c = 3.5$ hrs, where t_c was defined as the curing time of the epoxy phase before the polyurea was applied. Shorter t_c technically imply longer reaction times between the epoxy and polyurea prepolymers, i.e., a shorter epoxy cure time. A richer discussion of the influence of t_c on the foregoing properties follows. Zhou et al. [19] have shown indicate that the interfacial bonding properties between the two polymeric matrix phases, as well as the internal damping of the CHMC material, increase as t_c decreases. The study on the performance of CHMC wood stud walls subjected to tornado windborne debris impact revealed the influence of material impact resistance on t_c , where the material impact resistance increased as t_c decreased. The trend lines in Figure 9, which were obtained via linear regression, represent the crashworthiness parameters of specimens against different t_c values (see the dash and dash-dot lines), where the peak and average crushing forces increase as t_c decreases, where the differences are more discernable for the initial impact Peak Forces, F_{max} , as h_p increases. The higher load carrying capacity may be attributable to two possible mechanisms: (1) existence of an optimum reaction time between the epoxy and polyurea that leads to a higher interfacial bonding strength and toughness between the two polymeric matrix

phases, precluding delamination between the two polymeric phases; and (2) larger material damping as a result of the lower crosslinking density at the polymeric interface, effectively increasing load resistance at higher loading rates. In this latter case, molecular mobility and relaxation in internal energy dissipation, resulting in higher load capacity. Finally, it is believed that at slower loading rates (2 - 5 m/s), shorter t_c lead to an increase in F_{peak} (Figure 9a), yet as h_p increases, load recovery (F_{avg}) remains constant (Figure 9b) and the percentage of recovered load (F_{avg}/F_{peak}) actually decreases (from 62.9% to 37.5%). Longer t_c , on the other hand ($t_c = 3.5$ hrs), enable slightly greater percent load recovery (46.7% to 50.2%), a moderate increase in F_{max} and an increase in load recovery as h_p increases.

4.2.2. Energy dissipation/absorbing capabilities

The energy dissipation/absorption capability of materials is characterized by the total energy absorption (EA) and specific energy absorption (SEA). Similar to the load carrying capacity, the total energy absorption (EA) of CHMC specimens tends to increase as h_p increases and t_c decreases, a direct result of the increasing post-crushing average force values (Figure 9b), where EA is computed as the total area under the load-displacement curve. The EA and SEA are also larger for t_c equal to about 2.5 hrs as opposed to 3.5 hrs which appears to be consistent with the relationship of the load carrying capacities of the materials. An interesting observation is that the SEA, or energy absorption per unit mass of the crushed specimen, of CHMC specimens decreases with increasing h_p values. This may be a result of the density of the CHMC via inclusion of the hyperelastic polyurea matrix phase. As a result, even though the total energy absorption (EA) is observed to increase monotonically with increasing h_p values and

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4 increase with slower t_c ; and for certain fiber layouts and t_c values, there may exist an
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6 optimal h_p , t_c at which SEA peaks.
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9 The loading rate (or stroke traveling speed) at which the crashworthiness
10 parameters were obtained, was denoted as the numeric marks in Figure 9. At low impact
11 speeds (ranging from 2 m/s to 5 m/s), the effect of the loading rate appears negligible,
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13 and consequently, the change in strain rate in the constituent materials - as a result of the
14 different loading speeds - was not significant enough to cause noticeable difference in the
15 stress-strain responses.
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23 24 25 **5. Damage/failure mechanisms**

26 27 **5.1. Fractography of crushed specimens**

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29 The damage and failure mechanisms of both the CFRP and CHMC materials were
30 examined using scanning electron microscopy (SEM). A unique property of composite
31 materials is their evolutionary failure characteristic, where the non-homogeneity of the
32 microstructure provides numerous load paths around the damaged region. Because of the
33 non-homogeneity of the microstructure of composites, the integrity and response of the
34 composite are affected more by the effective accumulated subcritical damage than by any
35 single damage event. For the fibrous composites at sub-lamina level, this subcritical
36 damage may be manifested as matrix cracking, fiber-matrix debonding, fiber fractures,
37 etc. For instance, in the case of a unidirectional fiber reinforced composite under tensile
38 forces, the carbon fibers do not all snap at once in one fracture plane but instead, snap in
39 a sequential manner because of variability in flaw size along the fiber length. Thus, in
40 the design of a damage-tolerant composite material, "damage barriers" having a high
41 fracture toughness may be incorporated into a materials system such that any single
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6 characteristics as observed in many composites.
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9 The fracture surface morphologies of both carbon-fiber reinforced epoxy and
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11 CHMC coupons are presented in Figure 10. Because of the low fracture toughness of
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13 epoxy-based polymers and because the reinforcing fibers were orientated only in the
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15 longitudinal direction, cracks that had initiated at the weak fiber-epoxy interface quickly
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17 propagated throughout the CF/epoxy specimen, causing the CF/epoxy laminate to peel
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19 off the polystyrene foam core, see Figure 10a. The SEM image presented in Figure 10a
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21 also shows matrix cracking in the parallel-to-fiber direction. On the contrary, the CHMC
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23 laminates fail in a ductile manner. By introducing high fracture toughness and damping
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25 via the polyurea-epoxy interface and the polyurea matrix, damage paths are blunted and
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27 turned away, see Figure 10b, thus dissipating energy and preventing stress
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29 concentrations, where: (1) the crack does not readily propagate due to internal stress/
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31 force redistribution, (2) the crack location remains ‘localized,’ where the occurrence of
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33 the formation of similar cracks is reduced, and (3) otherwise common failure modes do
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35 not occur since damage events do not bridge together. Consequently, the CHMC system
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37 not only exhibits desirable macro-mechanical features, such as vibration suppression and
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39 high-impact resistance, but it also fails in a very ductile energy-releasing fashion with
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41 much significant damage tolerance.
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50 51 **5.2. Influence of the elapsed curing time of the epoxy phase - t_c**

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53 Additional discussion of the important t_c parameter is presented next. The
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55 crashworthiness of the CHMC specimen is largely influenced by t_c . Two possible
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57 mechanisms have been identified when the polyurea reacts with the epoxy resin [26-28].
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The first is a reaction between the polyurea end group (NCO) and the hydroxide group of the epoxy, see Figure 11a. A second mechanism may occur through the reaction between amine and an epoxide group, see Figure 11b [26]. The product of the polyurea-epoxy reaction is critical for the CHMC system: 1) the interface acts as a "compatibilizer" for improving the interfacial adhesion between the two polymeric matrix phases, which is critical for withstanding impacts and maintaining structural integrity; and 2) it provides additional source of internal damping through intermolecular friction or rotational mechanisms [33]. Furthermore, studies by Ma et al. [26;27] indicated that the morphology of the polyurea-epoxy reaction product is a function of the reaction time, where transmission electron microscopy (TEM) analysis reveals that with a decrease in t_c , i.e., an increase in the reaction time, the width of the polyurea/ epoxy interface increases, leading to a larger degree of crosslinked density of the reactive products. In that same study by Ma et al., the fracture toughness of the polyurea/ epoxy composite polymer was also shown to increase as the reaction time increases [26]. The "interruption" of the epoxy crosslinking reaction at an earlier stage (shorter t_c) produces a more compliant interfacial covalent molecular bond within the hybridized polymeric structure, resulting in higher mobility of the polymeric molecular chain. This consequently enables the internal friction between the molecules or the molecular rotations to produce higher damping and a larger molecular rotation component to exist [33].

6. Conclusions

The crashworthiness characteristics of a carbon-fiber reinforced polyurea-epoxy hybrid matrix composite were investigated through both quasi-static and low-speed impact crush tests. The methodology used to characterize the crashworthiness properties of composite

materials is discussed. The experimental test results on both the CHMC and conventional carbon fiber reinforced epoxies (CF/ epoxy) were presented, and the mechanism by which CHMC was able to sustain a high level of strength and to dissipate significant energy was discussed. The following conclusions are drawn from this study:

- The electron scanning microscopy (SEM) images revealed the multilayered texture of the polyurea-epoxy hybrid matrix system. A chemically well-bonded interface was developed between the epoxy and polyurea phases, and the morphology of which was shown to be a function of the elapsed curing time - t_c ;
- The comparison of quasi-static crush test results of CHMC with those obtained for the carbon-fiber reinforced epoxy (CF/epoxy) indicate that, as compared to traditional CF/epoxy, CHMC was able to maintain a relatively high post-peak strength and good structural integrity, precluding debonding of the Styrofoam core during post-crush stage. The energy absorption of CHMC was also substantially higher than that of CF/epoxy.
- The impact crush test results obtained for CF/epoxy and various types of CHMCs were presented. The influence of the two material processing parameters - h_p and t_c - on the crashworthiness parameters was investigated, and the results reveal that, generally, the load carrying capacity (F_{max} and F_{ave}) and total energy absorption (EA) increase with greater h_p and shorter t_c . However, the specific energy absorption (SEA) may reach an optimal state at certain h_p values.
- Examination of the fracture surfaces of both CHMC and CF/epoxy revealed that the inclusion of polyurea and the polyurea-epoxy interface following certain epoxy curing times in the hybrid matrix system had isolated the cracks, and successfully

mitigated certain damage modes, such as matrix cracking in the parallel-to-fiber direction. As a result, the damage tolerance of the material significantly improves under low speed impact crushing.

- Future research will focus on the effects of higher loading rates (in-depth) and the impact of the material processing on the properties of the CHMC as a function of parameters t_c and h_p . Future work will also focus on property identification using other fiber lay-ups, e.g., cross-ply, and load-bearing composites, e.g., glass-fibers, relative to the dual-hybridized epoxy-polyurea system, pertinent to automobile applications and more specifically to environmental durability and material performance under elevated temperature.

Acknowledgements

This research was partially supported by the Department of Homeland Security through the Higher Education Research Experience (HERE) Program, and the Southeast Region Research Initiative (SERRI) at the Department of Energy's Oak Ridge National Laboratory (ORNL), DHS Project No. 90300. This research through the Oak Ridge National Laboratory's High Temperature Materials Laboratory User Program was sponsored by the U. S. Department of Energy, Office of Energy Efficiency and Renewable Energy, Vehicle Technologies Program.

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Table 1 Specimen details and primary test results

Spec #	Processing Param. ¹	Test Condition	F_{max} (kN)	F_{ave} (kN)	CFE	Δ (mm)	EA ² (J)	SEA (J/kg)	Failure Mode ³
Quasi-static Tests (MTS810 - 7.6 mm/min)									
CF1s	$t_c=2.5\text{hrs}$ $h_p=1.59\text{mm}$	Quasi-static	10.427	4.423	0.424	63.5	297	--	LB, PF
CE1s	CF/epoxy		11.353	--	--	--	10.1	--	S
Impact Crush Tests (TMAC)									
CF2	$t_c=2.5\text{hrs}$ $h_p=2.29\text{mm}$	2m/s	22.969	8.438	0.367	148.5	1291	6598	LB, PF
CF3	$t_c=3.5\text{hrs}$ $h_p=1.37\text{mm}$		8.690	4.146	0.477	149.5	635	4751	LB, PF
CF4	$t_c=2.5\text{hrs}$ $h_p=2.94\text{mm}$		29.746	10.360	0.348	134.6	1449	6062	LB, PF
CF5	$t_c=3.5\text{hrs}$ $h_p=1.52\text{mm}$		12.600	5.479	0.435	161.8	911	6344	LB, PF
CF6	$t_c=2.5\text{hrs}$ $h_p=2.22\text{mm}$	3m/s	20.114	8.786	0.437	119.1	1099	5770	LB, PF
CF7	$t_c=3.5\text{hrs}$ $h_p=1.22\text{mm}$		9.478	3.844	0.406	129.6	523	4237	LB, PF
CF8	$t_c=2.5\text{hrs}$ $h_p=2.96\text{mm}$		28.327	10.513	0.371	117.7	1309	5460	LB, PF
CF9	$t_c=3.5\text{hrs}$ $h_p=1.64\text{mm}$		18.159	6.429	0.354	116.9	795	5245	LB, PF
CF10	$t_c=2.5\text{hrs}$ $h_p=1.69\text{mm}$	5m/s	15.994	11.203	0.700	115.2	1361	8799	LB, PF
CF11	$t_c=3.5\text{hrs}$ $h_p=1.18\text{mm}$		11.631	6.975	0.600	110.9	813	6735	LB, PF
CF12	$t_c=2.5\text{hrs}$ $h_p=2.80\text{mm}$		21.764	11.531	0.529	109.9	1347	5874	LB, PF
CF13	$t_c=3.5\text{hrs}$ $h_p=2.90\text{mm}$		16.321	8.968	0.549	112.4	1062	4497	LB, PF
CE2	CF/epoxy		13.164	8.965	0.681	22.5	240	2743	S

¹ t_c is the requisite curing time of the epoxy phase before applying the polyurea phase; smaller t_c values would result in longer reaction time between the epoxy and polyurea prepolymers. h_p is the nominal thickness of the polyurea layers as shown in Figure 2a, which is calculated via dividing the total weight of the polyurea phase by its density and total specimen surface area.

² The total energy absorption (EA) is calculated as the area under the load-displacement curves until the compaction stage was reached.

³ LB, PF: Local buckling and progressive folding; S: Splaying.

Figure 1. Load-displacement obtained from crush test

Figure 2. The dimension and material details of the CHMC rectangular tube specimen: (a) the texture of CHMC material; (b) SEM image showing the reinforcing carbon fiber and the dual-phase hybrid matrix system; (c) reinforcing carbon fiber; (d) zoomed-up view showing the interface between polyurea and BPA epoxy; (e) the dimension and layup of CHMC tube specimen

Figure 3. The test machine for automotive crashworthiness (TMAC) facility used for the experimental study: (a) overview; (b) close-up view of the test chamber; (c) specimen before crush (image captured by the Photron FASTCAM SA5 camera)

Figure 4. The load-displacement relation and energy absorption obtained from quasi-static tests

Figure 5. The load - displacement relationship and energy absorption of CHMC rectangular tubes under 2 m/s impacts

Figure 6. The load - displacement relationship and energy absorption of CHMC rectangular tubes under 3 m/s impacts

Figure 7. The load - displacement relationship and energy absorption of CHMC rectangular tubes under 5 m/s impacts

Figure 8. Crushing process and failure modes recoded by high-speed camera: (a) CHMC with $t_c=2.5$ hr, $h_p=2.297$ mm; (b) CF/epoxy

Figure 9. The influence of h_p and t_c on the crashworthiness characteristics of materials (the numbers in the figure symbol indicate impact speed, in m/s)

Figure 10. The fracture surface morphology of (a) Carbon-fiber reinforced epoxy; and (b) CHMC

Figure 11. Schematics of the polyurea-epoxy reaction: (a) reaction between the polyurea end group (NCO) and the hydroxide group; (b) reaction between amine and an epoxide group

Figure. 1.

Figure. 2.

Figure. 3.

Figure. 4.

Figure. 5.

Figure. 6.

Figure. 7.

Figure. 8.

Figure. 9.

Figure. 10.

Figure. 11.