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FAILURE MODES IN PARTICLE FILLED PLASTIC MATRIX COMPOSITES

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Abstract
Among the challenges that particle-filled plastics composites (PFPCs) present is the complexity of their mechanical behavior, particularly during plastic deformation. This challenge cannot be overlooked in need by most modern designs for materials with a combination of properties such as stiffness and strength than conventional plastics cannot meet. Thus, in this paper, detailed discussions on the influence of three failure modes, namely: matrix failure, failure of the particulate fillers, and interfacial decohesion on the mechanical stability of PFPCs, were done. The relationship between the matrix crack and the particulate filler was also discussed. At the same time, the mechanical behaviour of PFPCs when subjected to tensile, compressive, and impact loading was also reviewed. It is shown that the type of failure experienced by the particulate composite is influenced significantly by the mismatch in ductility and elastic modulus between the particulate fillers and the matrix. Factors such as temperature variation, increase in strain rate, presence of a sharp notch, formation of voids in the matrix, and poor adhesion between matrix and particle fillers with increased particle size also affect the failure mode experienced by the particulate composite.
Keywords: Composites, Failure, Fillers, Particulate, Plastics.

1. Introduction
Particulate fillers have been used in earlier studies individually in thermoplastics and thermosetting plastics [1-3]. Particle-filled plastics composites (PFPCs) are one of the cheapest and most widely used. PFPCs are subdivided into dispersion-strengthened and large-particle composites. The difference between these two subdivisions is based on reinforcement and strengthening mechanisms [4,5]. In dispersion-strengthened composites (DSCs), particle-matrix interactions that lead to strengthening occur on the atomic or molecular level, whereby the matrix bears the major portion of an applied load. At the same time, the small dispersed particles hinder the motion of dislocation (i.e., prevention of plastic dislocation such that yield, tensile strength, and hardness are improved). The dispersed particle diameter is between 0.01 and 0.1μm [4]. Large-particle reinforced composites (LPRCs) are composites to which fillers (discrete particles) have been added to modify or improve the properties of the matrix and replace some of the matrix volumes with less expensive material. The reinforcing particles have quite various geometries [4,6], but they should be of approximately equal dimensions in all directions (equiaxed). The particles are usually small (diameter greater than 0.1μm) and evenly distributed throughout the matrix for effective reinforcement. Furthermore, the volume fraction of the two phases influences the behaviour. At the same time, some mechanical properties are enhanced with increasing particulate content. The particle-matrix interactions cannot be treated on an atomic or molecular level; instead, continuum mechanics is used [4,6].

Two mathematical expressions Eq. (1) and (2), known as the rule of mixtures [4], have been formulated for the dependence of the elastic modulus on the volume fraction of the constituent phases for a two-phase composite. The equations predict that the elastic modulus should fall between an upper bound represented by;

$$E_c(u) = E_{mat}\, V_{mat} + E_{par}\, V_{par} \tag{1}$$

and a lower bound,

$$E_c(l) = \frac{E_{mat}\, V_{par}}{V_{mat}\, E_{par} + V_{par} E_{mat}} \tag{2}$$

E_c is the elastic modulus of composite, E_{mat} is the elastic modulus of the matrix, E_{par} is the elastic modulus of particle, V_{mat} is the volume fraction of matrix. At the same time, V_{par} is the volume fraction of particles. In order to extend the application area of plastics, reinforcing materials such as fibers, short fibers, or particles are added to the plastics matrix to produce plastics matrix composites (PMCs). These composites have become one of the new competitive materials in engineering. They are increasingly found in structural and industrial applications because of their high tensile strength, stiffness, and low density [7-9]. This work focuses on particle-filled plastic matrix composites.

2. Failure Modes

Most times, critical flaws and cracks in composite materials may go undetected [7]. In the case of particle reinforced composites subjected to mechanical loading, initiation, and progress of damage can occur through three basic local processes [10], namely, matrix failure, fracture of the particulate fillers, and interfacial decohesion as shown in Fig. 1.

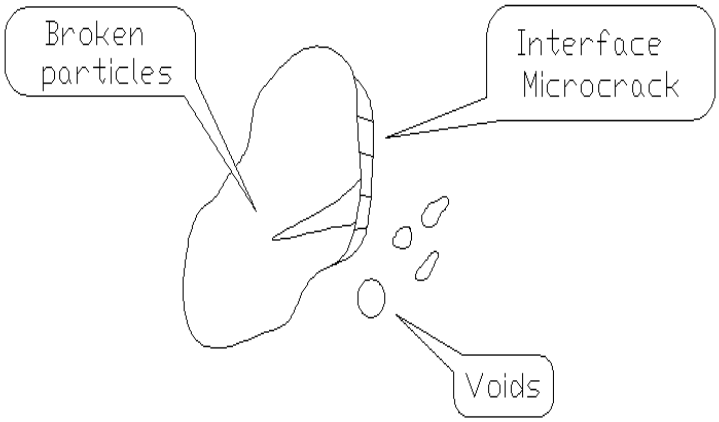


Fig. 1: Schematic diagram of different damage mechanisms in particle reinforced composites: particle breakage, interface debonding, and void growth in the matrix [11].

2.1. Matrix Failure

This is an important failure mode in thick rubber articles such as tire threads and tank track pads [12-15]. It has been discovered that when an elastomer (e.g., rubber) is subjected to repeated deformations with sufficient magnitude and frequency, considerable heat is generated, leading to a significant rise in temperature. Patel and Lee [16] and Soltani and Sourki [17] attributed this failure to the fact that though the addition of carbon black to rubber enhances mechanical and physical properties such as modulus, tear, tensile strength, and abrasion resistance; hysteresis properties such as heat build-up and resilience are being degraded depending on the sizes, shapes, and amount of the reinforcing particle. On the other hand, Glassy thermoplastics matrix could

experience a region of localized yielding, which leads to the formation of small and interconnected microvoids (see Fig. 2a). This phenomenon called crazing [4], which precedes fracture (see Fig. 2b), could occur due to factors such as reduction in temperature, increase in strain rate, and presence of sharp notch; in addition, it propagates perpendicularly to the applied tensile stress. Crazing could also be attributed to the entrapment of air in the matrix during the production of the composites even after complete solidification.

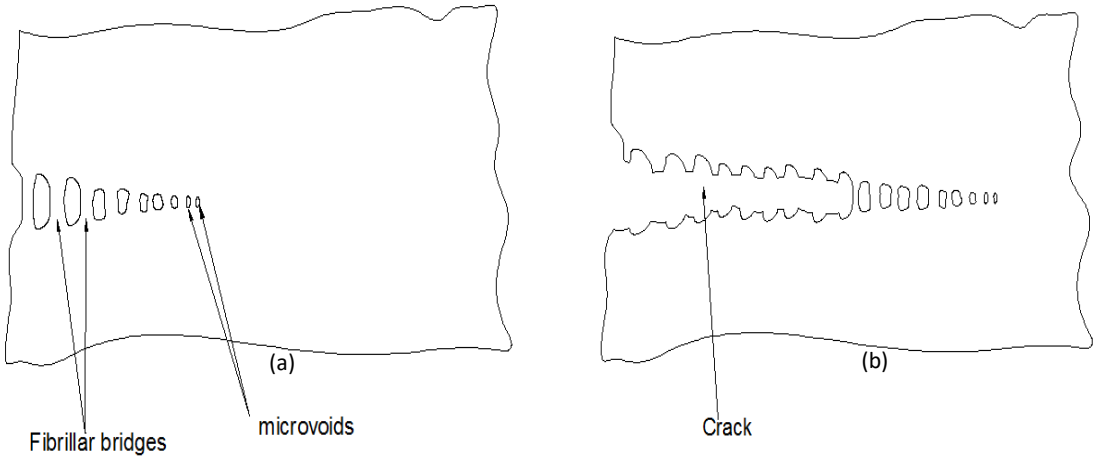


Fig. 2: Schematic drawings of (a) a craze showing microvoids and fibrillar bridges and (b) a craze followed by a crack [4]

This entrapped air is known as matrix porosity [18]. It is estimated by using the equations [18,19].

$$V_v = 1 - \frac{\rho_m}{\rho_{th}} \tag{3}$$

and

$$V_v = \rho_{mat} V_{mat} + \rho_{par} V_{par} \tag{4}$$

where, V_v is matrix porosity, ρ_m is measured density, ρ_{th} is theoretical density, ρ_{mat} is matrix density, ρ_{par} is particle density, V_{mat} is matrix volume fraction, while V_{par} is particle volume fraction.

The presence of porosity in the matrix structure has been undesirable. It weakens the composite and may lead to increased moisture absorption [19,20]. Various models have evolved over the years to effectively predict material failure due to void under different loading and environmental conditions [21-24]. The latter authors proposed the most popular void damage model currently used by the latter authors, the Gurson, Tvergaard, Needleman (GTN) model [24,25]. In 2000, Hao, Lui, and Chang (HLC) proposed a model [26], using the cell modeling technique (CMT), to simulate the effect of void nucleation, growth, and coalescence on the yield surface of a material. This model had been proven to describe the material behaviour of particulate composites appreciably. The CMT is useful and convenient because statistical averaging of the cell's constitutive behaviour over many "unit cells" can effectively represent the material's macroscale behavior [10,24,27]. Average stress and strain for a unit cell shown in Fig. 3 involve averaging the 'micro' stress and strain over the cell volume and applying the divergence theorem.

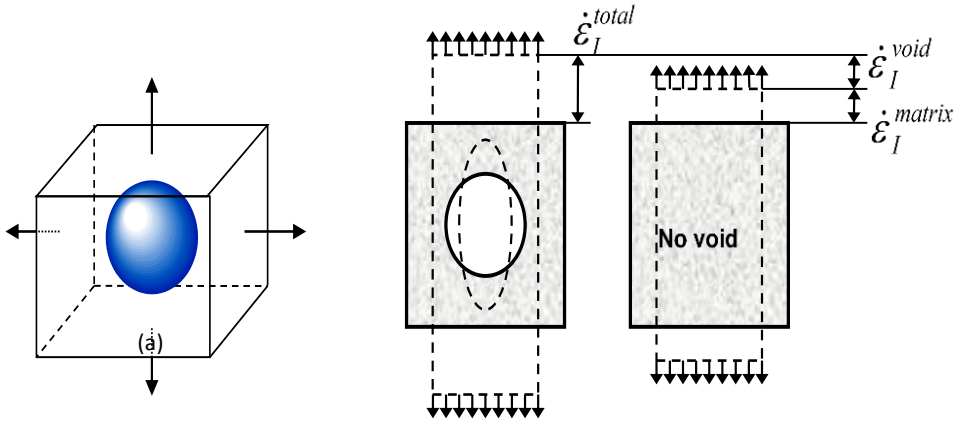


Fig. 3: Typical cell model, (a) with a central void, (b) additive behavior of strain rate due to voiding and strain rate in the matrix [24,26].

The resulting average stress and strain associated with a unit cell [24,26] is given as;

$$\varepsilon_{ij\,(average)} = \frac{1}{V_{cell}} \int_V \varepsilon_{ij}^{micro} dV = \frac{1}{2V_{cell}} \int_S \left(u_j n_i + u_i n_j\right) dS \tag{5}$$

$$\sigma_{ij\,(average)} = \frac{1}{V_{cell}} \int_V \sigma_{ij}^{micro} dV = \frac{1}{2V_{cell}} \int_S \sigma_{ik}^{micro} n_k x_j dS \tag{6}$$

where, ε_{ij} is Average strain in a cell, σ_{ij} is Average stress in a cell, V_{cell} is cell volume, ε_{ij}^{micro} is micro strain, σ_{ij}^{micro} is micro stress, u_j is material velocity components, n_i is unit normal to the cell surface, S is cell surface while x_j is spatial coordinates.

Decomposing the total strain rate in a cell-like that shown in Figure 4 into the strain rate due to the presence of a void and the strain rate in the matrix without damage, we have

$$\dot{\varepsilon}_I^{total} = \dot{\varepsilon}_I^{void} + \dot{\varepsilon}_I^{matrix} \tag{7}$$

where $\dot{\varepsilon}_I^{total}$ is total strain rate, $\dot{\varepsilon}_I^{void}$ is strain rate due to void damage, and $\dot{\varepsilon}_I^{matrix}$ is strain rate due to matrix material.

According to Hao *et al*. [26], during deformation of a unit cell, an instability point will be reached at which the deformation becomes unstable, and the ratio of the strain rate quantities, m_{20} is given as,

$$m_{20} = \frac{\dot{\varepsilon}_I^{void}}{\dot{\varepsilon}_I^{matrix}} \approx \frac{\Delta\varepsilon_I^{void}}{\Delta\varepsilon_I^{matrix}} \text{ when } \frac{d\sigma_I}{d\varepsilon_I^{matrix}} = 0 \tag{8}$$

The void volume fraction f_{void} is;

$$f_{void} = \frac{V_{void}}{V_{void} + V_{matrix}} \tag{9}$$

$$f_{void} \approx \frac{V_{void}}{V_{matrix}}; \quad \text{if } V_{matrix} >> V_{void} \tag{10}$$

where V_{matrix} is matrix volume, r is void radius, and V_{void} is void volume.

$$V_{void} = \frac{4}{3}\pi r^3 \tag{11}$$

2.2. Fracture of the Particulate Filler

Under uniaxial tensile loading, particulate fillers experience hydrostatic stresses, which could lead to brittle failure culminating in microscale damage along predefined "fracture surfaces" called equator region (the location where the interface normal is perpendicular to the loading direction) of the particle [10,27-30]. It has been experimentally established that in such materials, cracks tend to pass through the particles' centers, and such cracks are oriented normal to applied uniaxial load [10,29,31]. However, the transfer of the macroscopic stresses onto the particle is considerably more complicated [32]. This implies that a significant component of the particle stress is associated with the plastic flow of the surrounding ductile matrix, which is caused by macroscopic hydrostatic stress [32]. In some instances, macroscopic hydrostatic stresses can be transmitted through the matrix elastically. However, the elastic mismatch between the particles and the matrix leads to a concentrating effect of this pressure on the particles. Hence, it can be concluded that particle stresses are influenced by both the deviatoric and hydrostatic components of the macroscopic stress.

The radial displacement of the particle surface [33] is given by;

$$u_r = -\frac{r_p P_i}{3K_p} \tag{12}$$

where r_p is the radius of the particle, P_i is pressure on the particle, K_p is the bulk modulus of the particle; while the corresponding displacement of the matrix at the same location is;

$$u_r = \frac{r_p}{(1-f)}\left[\frac{f\,P_{in}-P_o}{3K_m}+\frac{P_{in}-P_o}{4G_m}\right] \tag{13}$$

$$f = \left(r_p \big/ b\right)^3 \tag{14}$$

where b is the radius of surrounding matrix shell, P_{in} is inner (hydrostatic) pressure on matrix, P_o is external pressure on matrix, K_m is the bulk modulus of the matrix, G_m is the shear modulus of the matrix, and f is particle volume ratio.

The particle experiences a different hydrostatic pressure P_i from the applied pressure because of the elastic mismatch between the particle and the matrix. Setting Eq. (12) and (13) equal and utilizing the usual relationship between constants, the pressure ratio is

$$\left.\begin{aligned} h\left(f,\upsilon_m,K_m/K_p\right) &= \frac{P_i}{P_o} \\ &= \frac{1+\left(\dfrac{2(1-2\upsilon_m)}{1+\upsilon_m}\right)}{1+\left(\dfrac{2(1-2\upsilon_m)}{1+\upsilon_m}\right)\left(f+(1-f)\dfrac{K_m}{K_p}\right)} \end{aligned}\right\} \tag{15}$$

2.3. Interfacial Decohesion

The interaction between matrix and particles (i.e., interface behaviour) is an important factor that influences particulate composites' mechanical properties [34]. Maiti and Singh [35] and Wall *et al.* [36] posited that the matrix and reinforcing particles, in the absence of coupling or dispersion agents, experience the formation of voids in the matrix and poor adhesion between matrix and reinforcing particles with increased particle size. Voids (air pockets in the matrix) are harmful because the matrix does not support the particles passing through the void. The addition of a coupling agent, also known as a bonding agent or binder, provides a flexible layer at the interface between particles and matrix that will improve their adhesion and reduce the number of voids trapped in the material [36]. Most analytical and numerical models assume that the bond between the particle and matrix is perfect and can be modeled using the continuity of tractions and displacements across discrete interface [37]. Though internal defects and imperfect interfaces exist in composites, incorporating them in the general theory requires modification and relaxation of the continuity of displacements between the constituents [38]. An imperfect interface bond may be caused deliberately by coating the particles. During the manufacturing process, it may also develop due to a chemical reactions between the contacting particle and the matrix material or due to interface damage from cyclic loading [37]. Basaran *et al.* [39] postulated that the strength of the interface bond controls the mechanical response and fatigue life of the composite.

3. Matrix Crack Interaction with Particulate Filler

Failure processes in dual-phase materials such as particle-filled composites are strongly linked to the basic problem of a matrix crack interacting with a second phase [40,41]. In an investigation into crack evolution in particulate composites (asphaltic mixture) by Soares *et al.* [42], the Cohesive Zone Model (CZM) developed by Barenblatt [43] was employed. The model assumes cohesive tractions acting in the fracture process zone ahead of the crack tips [43]. Considering a Cohesive Zone (CZ) in mode I fracture i.e., opening of the crack faces (Fig. 4), Barenblatt [43] postulated that separation commences at a theoretical tensile stress $\sigma_{\max}$, generally several orders of magnitude higher than the actual strength. Considering that stress transferring through the CZ depends on the relative displacements of the CZ faces, the constitutive relation $\sigma(w)$ between the cohesive stress σ and the separation distance *w* has to be a material property [44,45].

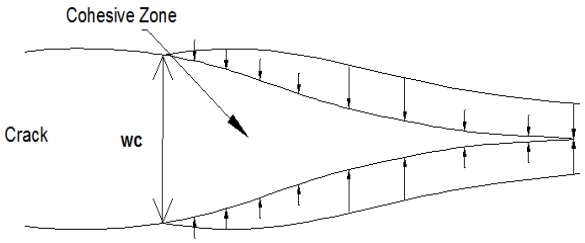
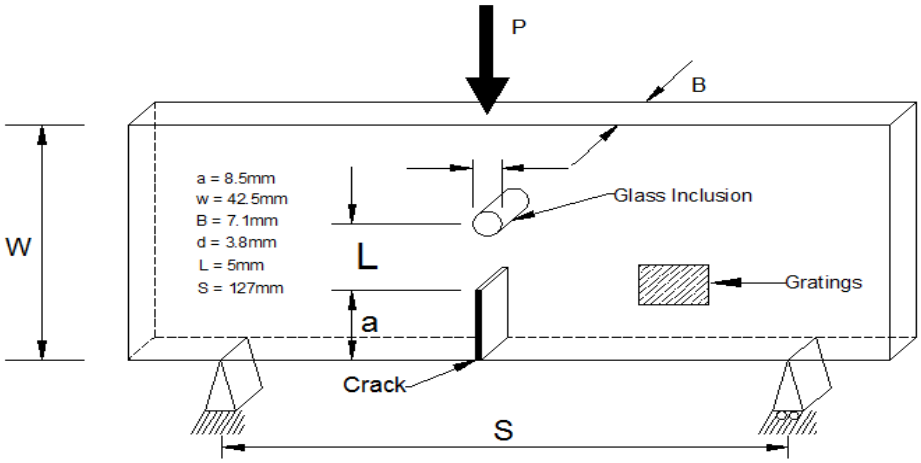
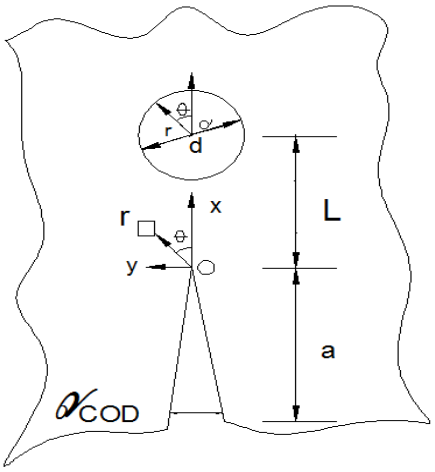


Fig. 4: Cohesive zone in mode I fracture [43]

Savalia *et al.* [40,41] considered a second phase inclusion in the composite matrix and the effect of the evolution of its debonding on crack tip parameters. Fig. 5 shows the edge cracked epoxy beam (with asymmetrically positioned cylindrical glass inclusion ahead of the tip) adopted in the investigation [40,41].



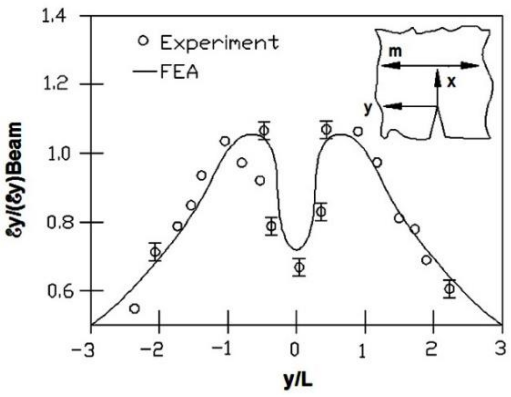
(a)



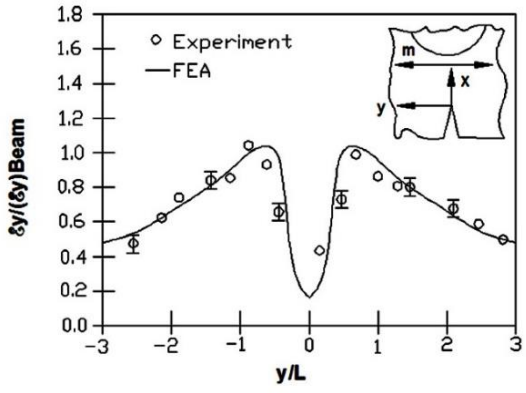
(b)

Fig. 5: Specimen details: (a) specimen geometry and loading configuration and (b) crack tip and inclusion [40,41].

They found that (Fig. 6a-c) substantial redistribution of strains ahead of the crack tip exists following the onset of debonding of the inclusion from the matrix. After debonding occurs, the strains close to $y = 0$ increase drastically and attain a peak value.



(a)



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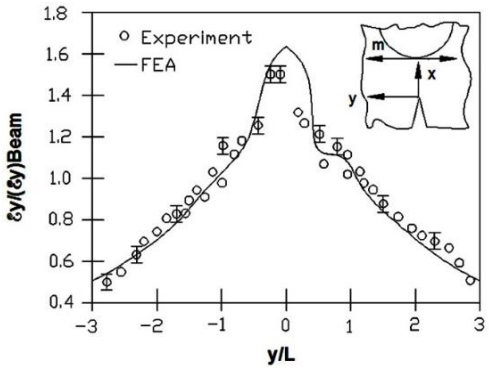

(c)

Fig. 6: Strain field evolution along (a) without inclusion (b) pre-debonding (with inclusion) and (c) post-debonding (with inclusion) stages [40,41].

4. Tensile, Compression and Impact Behaviour of Particulate Composites

The uniaxial tension test is the most fundamental method for the determination of data such as material specification, screening, research and development, and design of structural components [46-48]. Birley *et al.* [49], Starita [50], Pearce [51], and Deiter [52] also highlighted the importance of tensile behaviour of plastics matrix materials (PMMs) in terms of performance requirements. According to them, the key performance requirements for PMMs are tensile strength, compression strength, and impact resistance, while friction, wear, vibration, and damping may be of relevance in particular applications [51].

The work of El-hakim *et al.* [44] on the effect of surface treatment on the mechanical properties of polypropylene-bentonite composites showed that the treatment of bentonite acetyl trimethyl ammonium bromide (CTAB) caused more deterioration to the mechanical properties relative to oleic acid. In contrast, treatment with Tween 60 helped to keep the tensile yield unaffected (at 1% w/w) and delay the loss in the elongation up to 10pph loading of the filler (at 0.1% w/w), in comparison with the oleic acid treatment. A study on the application of bentonite as a filler in polypropylene composites by Othman *et al.* [46] shows that the value of the tensile strength, impact strength, and elongation at break decreases with increasing filler loading. Contrast increments in flexural and tensile modulus were observed as the loading of bentonite increases. The degradation temperature of composites decreases with increasing bentonite loading. In work by Grupta *et al.* [53] on the mechanical properties of polypropylene (PP)/LLDPE-copolymer blends and its glass compositions, it was discovered that the addition of the filler increased in tensile, flexural strength, and modulus, especially at 30% inclusion, where a manifold increase was observed, compared to the decrease recorded for varying PP/LLDPE mix. Stark and Berger [54] researched the effects of species and particle size on the mechanical properties of polypropylene filled with wood flour (ponderosa pine, loblolly pine, maple, and oak). It was found that Izod impact strength, flexural and tensile modulus, and strength increased with increasing particle size. With increasing wood flour content, flexural and tensile modulus, density, and notched impact energy increased.

In contrast, flexural and tensile strength, tensile elongation, and unnotched impact energy decreased. Dobreva *et al.* [55] investigated polypropylene's microstructural and mechanical properties filled with wood flour modified with monochloroacetic acid. It was found that homogeneity increased with increased monochloroacetic acid, while Young's modulus improved with increased filler content. In another investigation by Hietala [56] on the effects of extrusion parameters, different screw configurations, raw materials, and raw material pre-treatments on the

mechanical and microstructure of produced polypropylene filled with wood particles with a high aspect ratio, it was discovered that wood chips could be used as raw material in a one-step manufacturing process of wood-polymer composites. Bhaskar *et al.* [57] evaluated the properties of wood plastic composite (WPCs) made from matrices of recycled polypropylene (rPP) with sawdust (Pine Wood flour) as filler. The WPCs were produced using compounding and injection molding with varying formulations based on the plastic-type (PP), plastic form (recycled and virgin), wood four content, and addition of maleated polypropylene (MAPP) as coupling agent. It was discovered that incorporation of MAPP coupling agent in composite formulation improved mechanical stability. In related work, Medupin *et al.* [58] carried out impact and microstructural examination on low-density Polyethylene reinforced with sawdust from African teak and sodium hydroxide as a binder. A compression molding process produced the composites. It was discovered that that 40% of wood waste was the optimum reinforcement for impact strength. Microstructural examination revealed small discontinuities and reasonably uniform distribution of wood particles in the polymer matrix. Zhang *et al.* [59] investigated the influences of filler size and content on the impact strength and tensile strength properties of Al_2O_3/high-density polyethylene (HDPE) composites. It was discovered that the thermal conductivity and tensile strength of the composites increased with decreasing particle size. At the same time, the SEM micrographs of the fracture surface showed that Al_2O_3 with a small particle size was generally more efficient for enhancing the impact strength. Tavman [60] carried out a tensile test on aluminum powder-filled high-density polyethylene composites to determine mechanical behavior for varying volume content of aluminum particles. It was discovered that tensile strength and elongation at break decreased with increasing aluminum particle content.

In contrast, the modulus of elasticity increased up to around 12% volume content of aluminum particles due to the introduction of discontinuities in the structure by the filler particle. In a study by Rozman *et al.* [61] on the mechanical and dimensional stability of Rice Husk (RH)–Glass Fiber (GF) hybrid polyester composites, it was discovered that GF imparted higher tensile, flexural, and impact properties. At the same time, the presence of RH in the matrix produced the composites with comparable tensile, flexural, and impact properties, especially in the middle range of RH: GF ratios.

4.1. Tensile Test

Basaran *et al.* [62] carried out a tensile test on poly-methyl methancrylate (PMMA) filled with a fine dispersion alumina trihydrate (ATH), using in situ observations, to determine the influence of interfacial bond strength between particle and the matrix on the failure mechanism of the composite. They discovered that macroscopic failure is initiated by (i) the clusters of the reinforcing particles because of the strong interfacial bonding between particles and matrix (Fig.7) and (ii) separation of filler agglomerates from the matrix due to weak interfacial bond strength.

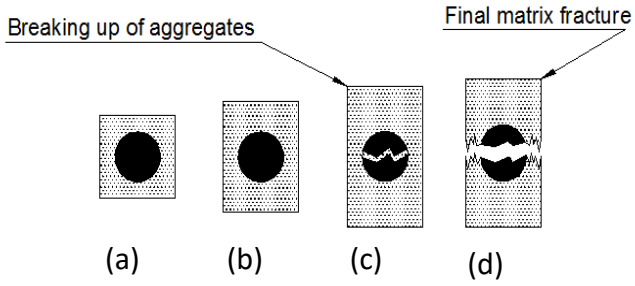

Fig. 7: Schematic of microdamage evolution in PMMA/ATH with a strong interfacial bond; (a) unloaded stage, (b) elastic deformation of matrix, (c) particle fracture, and (d) matrix and particle fracture [62].

According to Moshev and Evlampieva [63], the latter occurs in two stages: debonding of the matrix from filler and fracture of the matrix, as shown in Fig. 8.

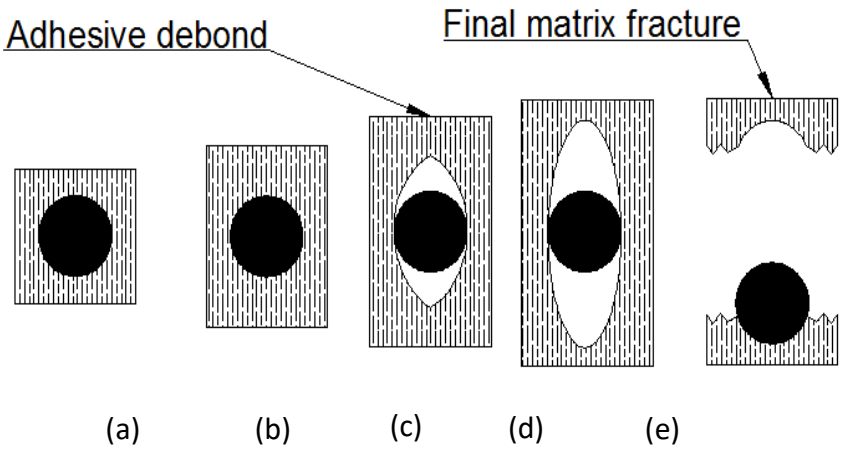

Fig. 8: Schematic of microdamage evolution in PMMA/ATH with a weak interfacial bond;

(a) unloaded stage, (b) elastic deformation of matrix, (c) adhesive debonding, (d) deboding of matrix from filler and (e) fracture of matrix [63].

4.2. Tensile Behaviour of Rigid Particle Reinforced Composite

The tensile behaviour of rigid particle-reinforced composites is influenced by the particle size, filler concentration, filler surface treatment, matrix and filler properties, superimposed pressure, and strain rate. It is well established that the fracture of particulate composites is associated with interfacial debonding between the matrix and particles, particle cracking, and the ductile plastic failure in the matrix depending on the relative stiffness and strength of the two constituent materials and the interface strength. If both constituent materials have material properties of the same order of magnitude or if the strength of the particle is low, particle cracking can occur. On the other hand, if the embedded particles are much stiffer and stronger than the matrix, matrix cracking (or cavity formation) and particle/matrix interface debonding become the major damage modes. These damage modes affect the important material properties such as tensile strength, fatigue strength, and fracture toughness. Most of the above discussions culled from Nie work [64] highlight the fact that emphasis should not be placed solely on the reinforcing effect of the particles and the influence of damage modes highlighted above on the mechanical performance of the composite.

Ravichandran and Liu [65] presented a schematic of a possible damage mode for a two-phase spherical particle reinforced composite (in perfect adhesion) subjected to tension (Fig. 9).

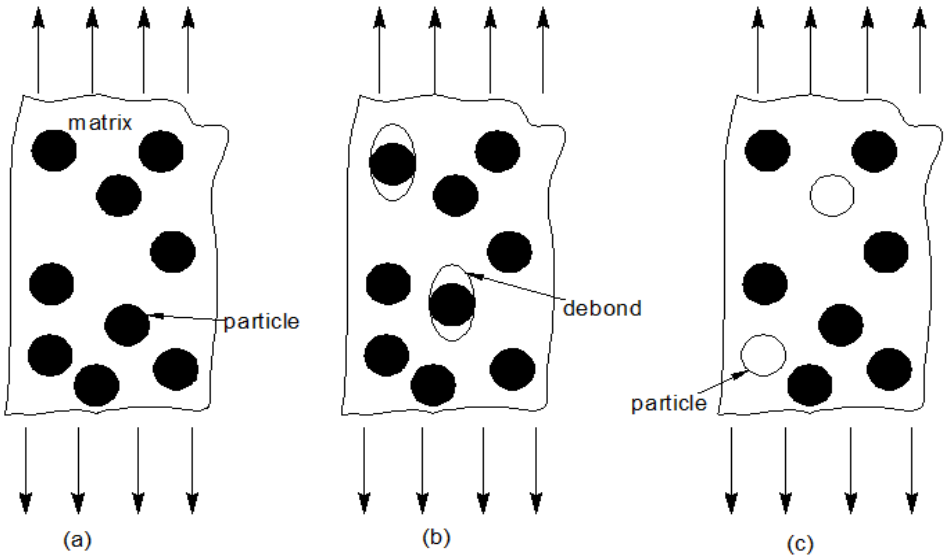

Fig. 9: Representation of particulate composite s subjected to uniaxial tension: (a) undamaged (b) damaged by debonding at the apex (c) damaged by cavity formation [67].

According to them, upon loading at a critical strain level, the matrix deforms more than the filler particle (interfacial debonding), as shown in Fig. 10(b). Then the formation of the cavity for well-bonded particles occurs Fig. 10(c). Tensile strength and modulus drastically decrease after debonding takes place, and there is a large increase in volume (dilation) as elongation continues [64,66,67]. In experimental work on the uniaxial tensile deformation of particle-filled polyolefin, Oshmyan and Muravin [68] presented two different modes of phase decohesion. According to them, at low filler fraction, as shown in Fig. 10(a), the micro homogeneous mechanism occurs when interparticle interactions may be neglected. It is characterized by complete uncorrelated debonding before yielding and similarity of further deformation in the vicinities of different particles. Above certain filler content (about volume fraction 20%), when interparticle interactions are noticeable, a micro homogeneous mechanism is changed to a craze-like process of which the characteristic features are correlated debonding in narrow zones, transverse to the loading direction, and concentration of further deformation mostly inside these microporous regions as shown in Fig. 10(b). It was discovered that as the concentration of filler increases, the elongation at which debonding takes place decreases. Hence, debonding mechanism is very important for further plastic deformation and fracture [64,68]. It determines ductile or brittle fracture, macro homogeneous flow, or deformation with necking.

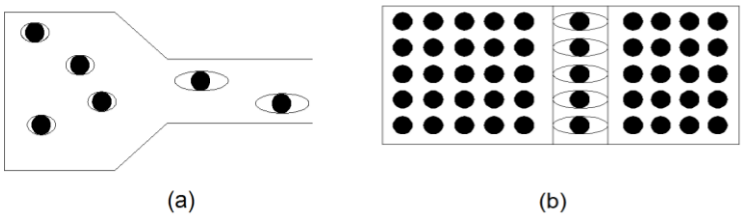

Fig. 10: Schematic representation of (a) micro homogeneous (b) craze-like deformation [68]

4.3. Determination of Effective Tension Properties

Determination of effective properties of composites is an essential problem in many engineering applications [69,70]. The properties are influenced by the reinforcement's size, shape, properties, and spatial distributions [65,66,69]. The rule of mixture, which is the simplest method leading to

the homogenized moduli of composite materials, was discovered by Kouznetsova *et al.* [70] in their work to be justified only for linear material properties. Eshelby [71] introduced a more sophisticated method called the self-consistent effective medium approximation, which was further developed by several authors [71-74]. In this method, equivalent material properties are derived from an analytical (or semi-analytical) solution of a boundary value problem for a spherical or ellipsoidal particle of one material in an infinite matrix of another material. Though the approach gives a reasonable approximation for structures that possess geometrical regularity, it fails to describe the behavior of clustered structures and high contrasts between properties of the phases [70]. Bensoussan *et al.* [75] and Sanchez-Palencia [76] developed a mathematical approach called the asymptotic homogenization theory. The method applies an asymptotic expansion of displacement and stress fields on the "natural length parameter," which is the ratio of a characteristic size of the heterogeneities to a measure of the macrostructure, to approximate their respective macroscopic distribution and then utilize variational principle to create a link between the scales [70,77-81]. This approach provides effective overall properties and local stress and strain values. Still, it is restricted to very simple microscopic geometries and small strains [63]. The unit cell methods (UCM), based on the concept of a representative volume element (RVE) originally introduced by Hill [73,81], became widely used [72, 83-90]. The homogenized material properties are determined by fitting the results of the detailed modeling of the RVE (typically performed by the FEM) on macroscopic phenomenological equations [73]. The UCMs account for complex microstructure morphology and investigate the influence of different geometrical features on the overall response. However, these methods do not provide adequate results for large deformations and non-linear history-dependent constitutive behaviour [73,90].

5. Compression Test

The compression test is conducted like a tensile test, except that the force is compressive (i.e., crushing load), and the specimen contracts along the stress direction. Compressive tests are used when the plastic flow behaviour and ductile fracture limits of materials under large and permanent (i.e., plastic) strain are desired, as in manufacturing applications, or when the material is brittle in tension [4,90]. According to the ASM Handbook [92], large L/D ratios should be avoided to prevent buckling and shearing modes of deformation. Figure 11 shows (a) Buckling when L/D > 5.0 (b) Shearing, when L/D > 2.5; (c) Double barreling, when L/D > 2.0 and friction is present at the contact surfaces; (d) Barreling when L/D < 2.0 and friction are present at the contact surfaces (e) Homogenous compression when L/D < 2.0 and no friction is present at the contact surfaces; (f) Compressive instability due to work-softening material.

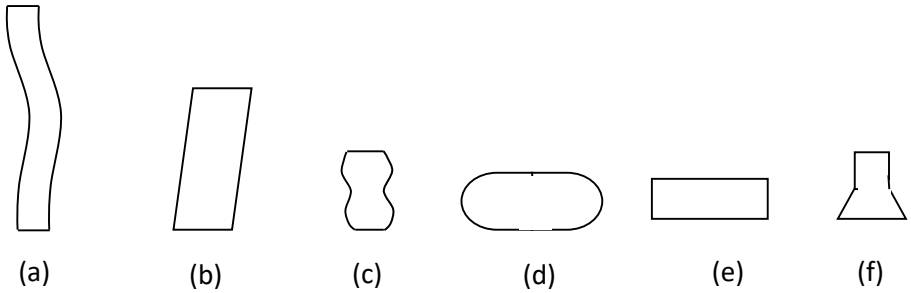

Fig. 11: ASM Handbook representation for a mode of compressive deformation [92]

5.1. Compressive Behaviour of Rigid Particle Reinforced Composites

The compressive behaviour of rigid particle reinforced composites is influenced by the same factors as in tensile conditions. When initially in compression, the particulate composite experience voids nucleation as shown in Fig. 12(a) at the particle-matrix interphase upon yielding [25].

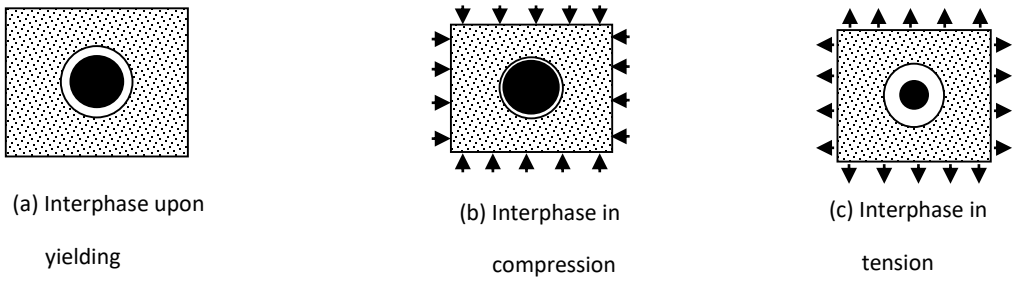

Fig.12: The influence of the second phase particle on void damage [25].

At this stage, the material, as a whole, behaves as if fully dense as shown in Fig. 12(b) with no voids (i.e., effective void volume fraction, $f = 0$), while in tension, the voids grow as shown in Fig. 12(c).

6. Impact Test

According to Galli [93] and Onitiri and Adeniyi [94], impact strength which is the ability to resist high-rate loading was discovered to be the most critical mechanical property of plastics in applications such as industrial, sports and aviation. This is because it (i) relates to the service life of the part, (ii) involve the increasingly important matters of product safety and liability, and (iii) is the first property to deteriorate as a result of environmental assaults such as temperature, UV exposure, chemical, etc. The studies of Howsz [95] also showed that impact resistance is a complex phenomenon that is sensitive to many testing variables such as time, temperature, and environment.

6.1. Impact Behaviour of Particulate Plastics Composites

Thio *et al*. [96] carried out a notched Izod impact test on polypropylene filled with calcium carbonate ($CaCO_3$) particles to study the mechanisms of deformation and fracture of isotactic polypropylene. Three types of particles with average diameters of 0.07, 0.7, and 3.5 μm were used at filler volume fraction from 0.05 to 0.30. It was discovered that the particles had either adverse or no effect on the impact toughness (due to poor dispersion) except the 0.7 μm diameter particles, which improved Izod impact energy up to four times that of the unfilled matrix. The toughening mechanisms were found to be plastic deformation of interparticle ligaments following particle-matrix debonding. Suwanprateeb [97] carried out notched Izod impact tests on CaCO3/high-density polyethylene composites with calcium carbonate in the range of 0-40% in a related work. He found that impact resistance decreased with increasing calcium carbonate. Also, transition in total impact energy and initiation energy with filler content was observed at 20% filler volume fraction (similar to propagation energy) and 10% filler volume fraction, respectively. The impact fracture process of Polyethylene was seen to be initiated by crazing, followed by brittle failure. In contrast, composites were initiated by microvoid nucleation from interfacial failure between the filler and matrix and were propagated via microvoid coalescence. Adedayo and Onitiri [98]

conducted notched Izod impact tests on iron ore tailings of particle sizes 150, 212, and 300 µm in polypropylene at volume fractions of 5 % to 30 %. It was discovered (Fig. 13) that after the initial drop at 5%, the addition of 300 µm iron ore tailings cause an increase in impact strength with increasing filler content contrary to the trend highlighted in Maiti and Mahapatro's work [99,100] on nickel-powder-filled PP and CaCO3 filled PP.

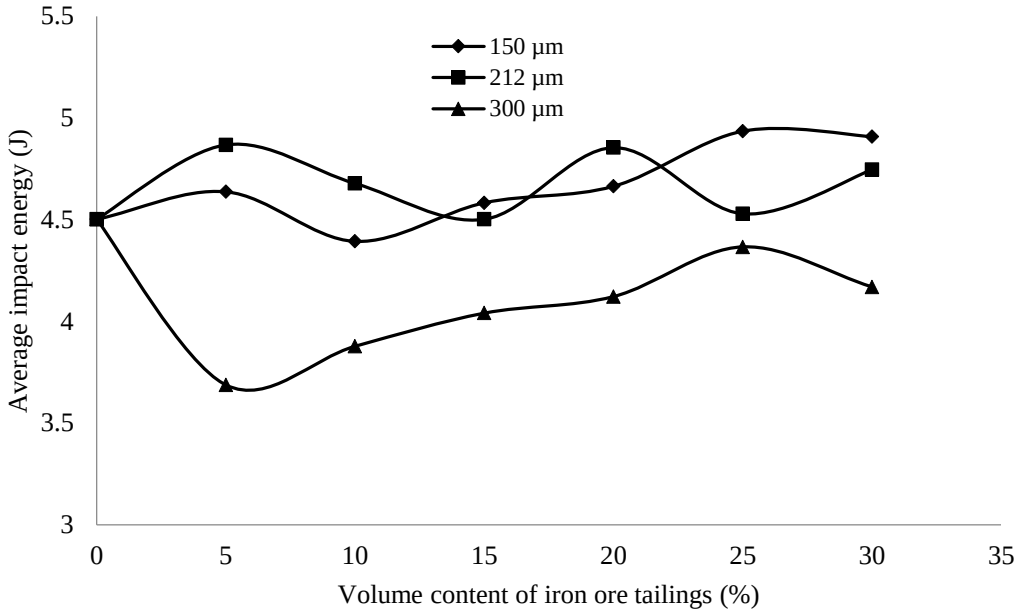

Fig. 13: Average impact energy versus volume content of iron ore tailings curves for PP-filled with 150µm, 212µm and 300µm iron ore tailings particle sizes at volume content 0 to 30% [98].

Arencón and Velasco [101] reported on Shelesh-Nezhad and Taghizadeh [102] and Maiti *et al.* [103] findings on notched impact tests on talc reinforced polypropylene composites. The former reported maximum impact strength at a talc content of 20 wt.% whereas the latter showed a dramatic decrease to 60% of the initial value of the PP matrix at a talc content of 30% volume.

7. Conclusion

A critical review of the effect of matrix failure, failure of the particulate fillers, and interfacial decohesion on the initiation and progression of damage in particle-filled composites subjected to mechanical loading is presented. It is shown that the type of failure experienced by the particulate composite is influenced significantly by the mismatch in ductility and elastic modulus between the particulate fillers and the matrix. Factors such as temperature variation, increase in strain rate, presence of a sharp notch, formation of voids in the matrix, and poor adhesion between matrix and particle fillers with increased particle size also affect the failure mode experienced by the particulate composite. Interfacial decohesion and formation of voids can be reduced considerably by the addition of a coupling agent, also known as a bonding agent or binder.

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