CHAPTER 4

RESULTS AND DISCUSSIONS

4.1. Determination of physical and thermal properties

4.1.1 Fibre volume fraction, V_f

Equations in section 3.4.2 (Chapter 3) were used to calculate the fibre volume fraction in these composites. It is common practice to use the resin density as the matrix density (PA6,6 = 1140 kg/m^3). Nevertheless, there is always a strong possibility that the resin has been modified by the presence of the fibres and that density of the matrix may be different from that of the resin due to nucleating effect of the fibres, molecular conformation effects of polymer chains at an interface, dissolution effects and reaction of the sizing [56]. However, when the tensile strength and tensile modulus are examined as a function of volume fraction, it can be seen in Table 4.21that the composite strength and composite modulus increase with the increase in the fibre loading.

Table 4.1 shows the experimental value of V_f and the values obtained are as expected, as physical blending approach had been adopted in mixing the compounded materials with the virgin matrix in order to reduce V_f of final composites. As reported by Hassan *et al.* [57], a precaution should be taken in preparing the materials before feeding into injection moulding. At the same fibre loading, the compounds were prepared in batches with a total weight of 500 g each. If large quantity of blend is used, the compounded materials tend to settle down in the feed hopper of the injection moulding machine, and leaving the polymer matrix at the top.

On the other hand, the same phenomenon was observed when mixing the glass fibre composites and carbon fibre composites to form hybrid fibre composites. This is due to the low density of carbon fibre. Therefore, the volume fraction of moulded specimen is not consistent.

4.1.2 Thermogravimetric analysis (TGA)

TGA was used to look at the thermal stability of the materials to ensure that the temperature chosen for drying and moulding did not lead to degradation. Figures 4.1 to 4.12 show the TGA thermograms for all composites in the range of study at different conditions.

The temperature range of 50-200°C corresponds to the gradual weight loss due to evaporation of water i.e., moisture. In dry samples, they did not show any existence of moisture at this temperature range because the samples were run immediately after moulding to get the dry as moulded condition. Whereas, 50% RH samples show some low molecular weight species burning off below 200°C, the percentage is less than 2% of the sample which may likely be due to the absorption of moisture in 50% RH conditions. For wet samples, the percentage of weight loss is higher i.e., 4% to 5% showing the maximum moisture absorption in the samples under the conditions employed.

The percent of weight loss in the temperature range of 300-500^{o}C was quite close to the matrix content of the composites (Figures 4.6 and 4.12). However, this is not an accurate technique to determine the fibre and matrix content in the composites since only a small portion of the sample was used during the test. Gradual weight loss between temperature range of 510-650^{o}C indicates the existence of fibres in the composites. There was no further weight loss after 700^{o}C until the end of the temperature program at 900^{o}C but slight increase observed in Figures 4.6 and 4.8. This observation might be due to the baseline drift of the instrument or possibly slight oxidation may have occurred as increased in carbon content. The sample was reduced to carbon black residue for the carbon and hybrid composite samples while glass fibres remain for glass fibre composite samples at 900°C.

4.1.3 Differential scanning calorimetry (DSC)

4.1.3.1 Reinforced polyamide 6,6

The melting point, T_m is mainly related to the degree of hydrogen bonding between the chains, which depends on the density of polyamide groups. For example, as the length of aliphatic groups between the amide links increases, the T_m will drop rapidly (e.g. nylon 6,6 melt at 260ºC, compared to nylon 6,12 at 212ºC) [58].

Table 4.2 shows the DSC results and Figure 4.14 illustrate the DSC scans of unreinforced PA 6,6 and reinforced composites used in this study at dry condition. The values of melting temperature, T_m of 4%, 8%, 13% and 16% V_f are essentially the same as the unreinforced PA 6,6. This suggests that there is no significant change in the microstructure of the matrix with addition of glass fibre. A similar observation is reported by Mohd Ishak *et al.* [59] for short glass fibre reinforced poly(butylene terephthalate) (PBT) composites.

The enthalpy, ΔH_m of caloric processes was determined at heating rate 5ºC/min. The ΔH_m of unreinforced PA 6,6 at transition temperature was 72.03 J/g, while the addition of glass fibre to the PA 6,6 system decreases the ΔH_m from 66.51 J/g to 50.30 J/g for composites with 4%, 8%, 13% and 16% V_f. This indicates that glass fibre absorbed more heat energy in melting the composites.

The crystalline peak temperatures T_c are also shown in Table 4.2. The T_c of unreinforced PA 6,6 is 233.7ºC and the addition of glass fibre increases the T_c to 236.1ºC. Glass fibre acts as nucleating agents which increase the crystallisation temperature of the matrix. The crystalline enthalpy, $-\Delta H_m$ of unreinforced PA 6,6 is 44.52 J/g. As glass fibre is added, the $-\Delta H_m$ decreases. It is assumed that the decrease of $-\Delta H_m$ is closely related to the changes in crystal from triclinic to hexagonal form since the addition of glass fibre interfered with crystallisation. The crystallisation occurs because the glass fibre plays a role as a nucleating agent [60, 61].

The results of DSC are reported in Table 4.3 and the DSC heating scans are illustrated in Figure 4.14 for unreinforced PA 6,6 and reinforced composites conditioned at 50% RH. The similar trend is also observed in dry condition. Table 4.3 shows no changes in T_m and T_c while a decrease in ΔH_m and $-\Delta H_m$ is observed as the fibre is added to the PA 6,6 system.

By contrast, the T_m decreases with addition of fibre to polyamide system at wet condition as tabulated in Table 4.4 and illustrated by Figure 4.15. The T_m of unreinforced PA 6,6 is 262.9ºC and decreases to 260.7ºC when fibres are added up to 13% to the composite system. This phenomenon may be attributed to changes of crystal structure of PA 6,6 or variation of crystal thickness and size. Mohd Ishak *et al.* [59] reported a similar shifting of the T_m when the specimens were exposed to water. However, for 16% V_f composite the T_m is only 0.4ºC lower compared to unreinforced PA 6,6. The ΔH_m of unreinforced PA 6,6 at transition temperature is 66.1 J/g, while the addition of 4%, 8%, 13% and 16% glass fibre to the PA 6,6 system decreases the ΔH_m to 47.05 J/g.

The crystalline peak temperature T_c of unreinforced PA 6,6 is 236ºC and the addition of 13% glass fibre in the composite decreases the T_c to 233.9ºC. The moisture is expected to decrease the crystallisation temperature of the matrix which indicates that the existence of water in PA 6,6 system breaks up the intermolecular hydrogen bonding and allows the hydrocarbon chain segments to pack better [58]. The crystalline enthalpy, $-\Delta H_m$ of unreinforced PA 6,6 is 43.11 J/g. In general, as glass fibre content was increases, the $-\Delta H_m$ decreased. The clear trend of increasing and decreasing of these values with function of condition can be seen in Table 4.5.

4.1.3.2 Carbon fibre composite

The DSC heating scans are given in Figure 4.16 and the Table 4.6 represents the DSC results of carbon fibre composites at dry, 50% RH and wet conditions. Similar behaviour as glass fibre composites are observed as moisture absorption increase. No significant changes in the T_m of carbon fibre composites as moisture content increases. The ΔH_m of these composite at transition temperature is decreased from 34.58ºC to 34.29ºC when the moisture content increases. The changes in T_c value for these composites are too small or negligible as moisture content increases. Meanwhile, $-\Delta H_m$ show a decrease from 21.95ºC to 21.47ºC from dry to wet condition.

4.1.3.3 Hybrid fibre composite

Table 4.7 shows the DSC results and the DSC scans of glass/carbon hybrids fibre composites at dry condition are illustrated in Figure 4.17. The differences in T_m and T_c for single glass and single carbon fibre composites are too small or negligible, indicating no changes in the crystalline structure. ΔH_m values of single glass and single carbon fibre composites at transition temperature are 47.37 J/g and 34.58 J/g, respectively. The $-\Delta H_m$ value of glass and carbon fibre composite is 29.59 J/g and 21.95 J/g.

Table 4.7 also shows the hybrid effect with the hybrid composition. It can be seen that the values of ΔH_m and -ΔH_m reduce as the relative proportions of glass fibre are decreased. The decrease in ΔH_m value suggests that carbon fibre absorbs more heat energy in the melting of the composites as the ΔH_m of carbon fibre was much lower than single glass fibre composite. Similar behaviour is also observed in 50% RH and wet samples as for dry condition samples (Tables 4.8 and 4.9 and Figures 4.18 and 4.19).

Table 4.10 represents the DSC results of glass/carbon hybrid fibre composites subjected to dry, 50% RH and wet conditions. Similar behaviour for both glass fibre composites and carbon fibre composites are observed. The T_m of SG75/C25 composite is 258.7ºC and increases to 259.9ºC as the moisture content increases. The ΔH_m of these composites at transition temperature are decreased from 45.35 J/g to 42.14 J/g when the moisture content increases. The increment of crystalline peak temperatures, T_c to 233.8ºC and the decrement in -ΔH_m to 25.79 J/g are observed with respect to 50% RH and wet conditions. In contrast, the changes in T_m, ΔH_m, T_c and -ΔH_m of SG50/C50 and SG25/C75 composite are too small or negligible when exposed to 50% RH and wet condition.

4.1.4 Dynamic mechanical analysis (DMA)

4.1.4.1 Glass fibre composites

Dynamic mechanical curves of injection moulded composites are given in Figures 4.21 to 4.38 depicting tan δ and storage modulus, E' curves. Tan δ is the ratio of the loss component to the storage modulus. It can be seen from Figures 4.20 to 4.25, the tan δ and storage modulus curves were respectively shifted to lower and higher values by the introduction of fibre reinforcement. From the tan δ curves, over a temperature range of -100ºC to 200ºC, two transition regions as indicated by two damping maxima, are observed. These transitions are usually known as ß-transition at a lower temperature and α-transition at a higher temperature.

In dry condition (Figure 4.20), the temperatures at the maximum value for tan δ in ß-transition and α-transition for unreinforced PA 6,6 are recorded at -70ºC and 55ºC respectively. For the reinforced PA 6,6, the temperatures are in the range of -69ºC to -63ºC and 52ºC to 71ºC, respectively for the ß-transition and α-transition. Whereas for 50% RH sample (Figure 4.22), the temperatures at the maximum value of tan δ in ß-transition and α-transition for unreinforced nylon 6,6 are recorded at -74ºC and 17ºC respectively. For the reinforced PA 6,6, the temperatures are in the range of -76ºC to -74ºC and 11ºC to 19ºC, respectively for the ß-transition and α-transition. However in wet sample (Figure 4.24), only α-transition appears and the temperature at the maximum value of tan δ was recorded at -13ºC for unreinforced PA 6,6. The temperatures are in the range of -19ºC to -16ºC for reinforced PA 6,6 at α-transition. In wet or humid conditions, the absorption of water from these materials leads to plasticising effect of the matrix and causes a loss in mechanical performance [62].

From the tan δ curves, thermomechanical data are summarised in Tables 4.11 to 4.14. In these tables, tan δ_{25} and tan δ_{max} are the tan δ values at a temperature of 25ºC and the maximum value of the tan δ respectively. T_α and $T_ß$ are the temperatures at the maximum value of tan δ in α-transition and ß-transition respectively. The α-transition temperature is commonly known as the glass transition temperature, T_g [63]. Referring to the Table 4.11, tan δ_{25} in dry condition decreases by the introduction of fibre reinforcement. In the α-transition region, tan δ maxima, tan δ_{max} is decreased; glass transition, T_g, is increased as the fibre volume fraction increases. In the ß-transition region, tan δ_{max} is decreased and the temperature at a maximum value of tan δ, $T_ß$ shows a decrement as the fibre volume fraction increases.

Similar behaviour has been observed in 50% RH (Table 4.12) and wet conditions (Table 4.13) as the fibre content increases. Nevertheless, there are no values of tan δ_{max} and the temperature at a maximum value of tan δ, $T_ß$ at ß-transition region are recorded for wet sample. This is due to limited machine capability which can reach until -100ºC.

Dynamic mechanical curves of injection moulded composites at different conditions; i.e. dry, 50% RH and wet conditions for each volume fraction are given in Figures 4.26 to 4.37 respectively depicting tan δ and storage modulus, E' curves. It is clearly seen that the tan δ and storage modulus curves are respectively shifted to lower and higher values when expose to humid environment. The temperatures at the maximum value of tan δ in ß-transition and α-transition for unreinforced PA 6,6 and reinforced PA 6,6 at different conditions have been given.

From Table 4.14 tan δ_{25}, is the highest when exposed to 50% RH environment and lowest in dry condition. In the α-transition region, tan δ_{max} is increased; glass transition, T_g, is decreased when moisture absorption increases. It is shown that humidity acts as a plasticiser which induces an increase in the amorphous chain mobility in the material and hence reduces the glass transition temperature, T_g significantly [62].

In the ß-transition region, tan δ_{max} is increased when exposed to humid environment. The temperature at a maximum value of tan δ, $T_ß$ is slightly decreased when exposed to humid environment. However, no values are recorded for wet samples. All these parameters suggest that the higher content of moisture will introduce the plasticising effect in the composites which can lower the glass transition temperature, T_g upon exposure to moisture.

Aitken *et al.* [44] have reported the reduction in glass transition temperature, T_g from 90ºC to 57ºC when studying the effect of moisture on acrylic fibres. This shows the plasticising action of water on the samples when immersed in water at different temperatures.

4.1.4.2 Carbon fibre composites

Dynamic mechanical curves of injection moulded, carbon fibre composites are given in Figures 4.38 to 4.39, depicting tan δ and storage modulus, E' curves. Over a temperature range of -100ºC to 200ºC, two transition regions are recorded throughout composition ranges. For dry sample, the temperatures at the maximum value of tan δ in ß-transition and α-transition are recorded at -66ºC and 69ºC respectively. Whereas for 50% RH sample, the temperatures at the maximum value of tan δ in ß-transition and α-transition are recorded at -76ºC and 19ºC respectively. However, only α-transition appears in wet sample and the temperature at the maximum value of tan δ is recorded at -11ºC. The same data of the tan δ curves for the glass fibre composites are observed as shown in Table 4.15. Definitions of the symbols used are the same as mentioned above for the glass fibre composites.

The same behaviour as in the case of glass fibre composites is noted. The tan δ and storage modulus curves are respectively shifted to lower and higher values as the moisture absorption increases. Generally, the effect of conditioning these composites gives similar results as for the glass composites, except that the magnitude of change for carbon composites is bigger. This was expected due to the effect of more fibre fraction in carbon fibre composites.

4.1.4.3 Short carbon/short glass hybrid fibre composites

Figures 4.40 to 4.45 illustrate tan δ and storage modulus, E' curves for glass/carbon hybrid fibre composites. Over a temperature range of -100ºC to 200ºC, two transition regions are recorded throughout composition ranges. The temperatures at the maximum value of tan δ in ß-transition and α-transition are -66ºC and 66ºC to 67ºC, respectively in dry condition (Figure 4.40). Moreover, in 50% RH condition (Figure 4.42) the temperatures at the maximum value of tan δ in ß-transition and α-transition shifted to lower values where the temperature range are -77ºC to -75ºC and 16ºC to 18ºC, respectively. Nevertheless, when samples are immersed in water, i.e. wet condition (Figure 4.44), only the temperatures at the maximum value of α-transition are recorded in the range of -14ºC to -12ºC.

The same data of the tan δ curves for glass and carbon fibre composites are observed. Definitions of the symbols used are the same as mentioned above for the glass and carbon fibre composites.

Tables 4.16 to 4.18 present the thermomechanical data for these composites. In dry condition (Table 4.16), tan δ at a temperature of 25ºC, tan δ_{25}, shows an increment with increase of the carbon fibre ratio and corresponding decrease of glass fibre ratio. In the α-transition region, tan δ_{max} is higher in hybrid system compared to the single glass and single carbon composites. While glass transition, T_g, does not show much difference between single fibre composites and hybrid fibre composites. In the ß-transition region, tan δ_{max} decreases with increase of the carbon fibre content and corresponding decrease of glass fibre ratio. $T_ß$ shows a decrement in hybrid system compared to the glass and carbon composites.

In contrast tan δ_{25}, shows a decrement with increase of the carbon fibre ratio and corresponding decrease of glass fibre ratio in 50% RH condition (Table 4.17). Whereas, tan δ_{max} in the α-transition region is lower in hybrid fibre composite samples compared to the glass and carbon fibre composites. Glass transition, T_g, does not show much difference between single composites and hybrid composites. In the ß-transition region, tan δ_{max} shows similar behaviour as in dry condition. $T_ß$ only shows a small difference in both hybrid fibre composites and single fibre composites. The same behaviour as in the case of dry condition is noted in wet condition (Figure 4.17). From the observation, the tan δ and storage modulus curves are respectively shifted to lower and higher values with the addition of glass and carbon fibre at different ratios.

The effect of conditioning on glass and hybrid fibre composites is shown in Figures 4.46 to 4.51. It's clearly seen that the tan δ and storage modulus curves are respectively shifted to lower and higher values when moisture content increases. The temperatures at the maximum value of tan δ in ß-transition and α-transition for glass and hybrid fibre composites at different conditions have been reported. The same data of the tan δ curves for glass and carbon fibre composites are observed.

The same behaviour as in the case of single fibre composites is observed. The tan δ values are further shifted towards lower temperatures when samples are exposed to high humidity environment. From Table 4.19, it can be seen that generally the effect of conditioning is the same for single fibre composites, except the magnitude of the change is bigger. This is probably due to the effect of introducing two different fibres in the system.

4.2 Fibre length characteristics

4.2.1 Effect of processing on fibre length distribution

Processing of fibre glass into injection moulded composites leads to large reduction in the fibre length as reported by Thomason [56, 64]. The histograms of fibre length distribution (FLD) of glass fibre composites (Figure 4.52) and carbon fibre composites (Figure 4.53), which are presented in the form of the percentage of fibre count versus mid-point of fibre length ranges with 0.1 mm interval for each before and after moulding process.

It can be seen that majority of the glass fibres and carbon fibres have length 0.3 mm (mid-point fibre length is 0.25 mm) before moulding and the fibre length reduced to 0.2 mm after moulding process. This is due to the fact that processing of materials lead to fibre degradation. In the case of carbon fibre, the fibres are more brittle and thus fracture more easily than the glass fibres during processing.

4.2.2 Effect of fibre volume fraction on fibre length distribution

Figures 4.54 to 4.59 show the FLD of glass and carbon fibre composites which are presented in the form of the percentage of fibre count versus mid-point of fibre length ranges with 0.1 mm interval for each.

Figure 4.60 illustrates the effect of fibre volume fraction on FLD of glass fibre composites and Figure 4.61 shows the cumulative FLD of injection moulded glass fibre composites. Most of the fibres have lengths between 0.1 mm to 0.3 mm. Composites SG4, SG8 and SG13 have 30.1%, 35% and 31.4% of fibres respectively with lengths shorter than 0.3 mm compared to composite SG18 which has 41.6% of fibre length less than 0.3 mm. It can be seen that as fibre volume fraction increases, most of the fibres tend to degrade during processing. This type of behaviour can also be seen from the calculated values of the percentage of fibres with length, L less than 0.2 mm, 0.4 mm and 0.6 mm (Table 4.20). It can be seen that at any L below 0.6 mm, the composites have less percentage of fibres indicating that there are more fibres with L longer than 0.6 mm.

Fu *et al.* [65] reported similar finding and the authors quoted that the increased damage to the fibre length with increase in fibre loadings is attributed to the greater fibre-fibre interaction. Other authors [26, 66-68] also reported that fibre degradation is due to an increase in fibre volume fraction of composites.

4.2.3 The statistical analysis of fibre length distribution

The number average, L_n and the weight average L_w of fibre length are commonly used to summarize the properties of the fibre length distributions. L_n is expressed by

$$L_n = \frac{\sum_{i=1}^{N} n_i l_i}{\sum_{i=1}^{N} n_i} \tag{4.1}$$

whereas the L_w is expressed by (this expression is valid only if fibres of equal diameter are assumed)

$$L_w = \frac{\sum_{i=1}^{N} n_i l_i^2}{\sum_{i=1}^{N} n_i l_i} \tag{4.2}$$

where; n_i is the number of fibre count of fibres of species i with length l_i and N is the number of fibre length ranges (fractions). The number average reflects the level of fibre damage during processing whereas the weight average is likely more useful in describing the composite mechanical properties which relate to volume fractions.

The effect of fibre volume fractions on both L_n and L_w is presented in Figure 4.62. Both L_n and L_w decrease almost linearly with increasing fibre volume fraction. The same trend of behaviour was also reported by Thomason [56] and Hassan *et al.* [66]. In explaining this, the author quoted that increase in fibre volume fraction leads to increase in probability of fibre-fibre and fibre-machined interaction and an increased apparent melt viscosity resulting in higher bending forces on the fibres during compounding and moulding. Hassan *et al.* [66] in his study on short and long glass fibre reinforced PA 6,6 also reported that fibre-fibre, fibre-metal and fibre-matrix friction occur during compounding and moulding process which leads to fibre degradation.

4.3 Tensile properties

4.3.1 Single fibre composites

4.3.1.1 Glass fibre composites

Table 4.21 shows tensile properties such as tensile strength, tensile modulus and fracture strain for glass fibre composites and the histogram of these properties are illustrates in Figures 4.63 to 4.65. All specimens are hard and brittle except some indication of ductility shown by sample SGD0, the unreinforced PA 6,6. This sample indicates severe localised deformation (necking) before fracture. However, this behaviour is not observed in samples with fibre reinforcement.

It can be seen that increasing the fibre content results in an increase in composite strength over the range of the study. Both tensile strength and modulus of composites increase with fibre volume fraction. The tensile strength for unreinforced PA 6,6 in dry, 50% RH and wet conditions are found to be 53 MPa, 50 MPa and 35 MPa respectively. As the fibre volume fraction increases, 67-200%, 20-110% and 23-140% increments in tensile strengths at dry, 50% RH and wet condition are observed compared to unreinforced PA 6,6 counterparts.

Furthermore, the tensile moduli for unreinforced PA 6,6 in dry, 50% RH and wet conditions are found to be 0.71 GPa, 0.62 GPa and 0.38 GPa respectively. As the fibre volume fraction increases, 130-437%, 0-356% and 92-508% increments in tensile moduli at dry, 50% RH and wet condition are observed, compared to unreinforced PA 6,6 counterparts.

This is expected as increasing amount of glass fibre leads to high stiffness of composites and more energy is needed to break the composites. The applied load is transferred to the strong and stiff fibres through the fibre-matrix interface. As fibre volume fraction increases, more fibre-matrix interfacial area is created and the more applied load is transferred to the fibre via the interface. Thus, it is more difficult to break and hence results in greater tensile strength and tensile modulus. The composites in all fibre loadings at same condition show similar trend. A similar observation on tensile strength and tensile modulus has been reported by Mohd Ishak *et al.* [59] for short glass fibre reinforced poly(butylene terephthalate) (PBT) composites.

From Figure 4.171(b), it can be seen that the fibre population is more compared to Figure 4.171(a). This is relatively in agreement with the amount of fibre need during preparation. As expected, these results support the increment of tensile strength and modulus.

However, if the fibre-matrix adhesion is weak, the cracks tend to form at the interface and link up quickly through highly stressed sections of matrix, resulting in the premature failure of the composite. At high fibre loadings, there are more places to form interfacial cracks in the composite system with weak fibre-matrix adhesion. Nevertheless, when the tensile strength and tensile modulus are examined as a function of volume fraction, it can be seen that the composite strength and composite modulus increases with the increase of the fibre loading. This shows that fibre-matrix adhesion for these composites are strong. Thomason and Fu *et. al* [56,69] also reported the equivalent results and showed that the fibre-matrix interfacial adhesion strength increase with the PA 6,6 concentration in matrix system.

Tensile strength can also be influenced by fibre orientation. Random orientation of the short fibres result in lower reinforcement effectiveness in fibre reinforced compounds [70]. This is affected by processing parameter; melt flow direction (MFD). The fibre volume fraction may be controlled by the fabricator, within the limits allowed by the requirement for adequate flow during processing, but the fibre aspect ratio and orientation is largely dictated by the flow history of the material during processing of polymer [71]. During processing, injection moulded composites exhibit a complex distribution of fibre orientations due to the interactions between melt properties and moulding conditions. These general features are apparent in studies of fibre-orientation distribution found in the literature [56,72]. Figure 4.173 clearly shows the misalignment of the fibres in the composites at various magnifications. The misalignment of the fibres normally reduced the tensile strength of the composites compared to the unidirectional short fibre composites.

The fibre volume fraction has an evident influence on the composite failure strain. The fracture strain as a function of fibre volume fraction is shown in Figure 4.65. As seen in Table 4.21 the fracture strain for unreinforced PA 6,6 in dry, 50% RH and wet conditions are found to be 231%, 349% and 149% respectively. As the fibre volume fraction increases, 96-97%, 91-97% and 80-94% decrements in fracture strains at dry, 50% RH and wet condition are observed compared to unreinforced PA 6,6 counterparts. Results obtained were not as expected. The inconsistency in the results could possibly be attributed to the experimental error as measurements were made. If the extensometer is functioning, it will give consistency in readings.

Most of the previous studies [73-76] also reported the decrement in fracture strain as the fibre volume fraction increases. Due to the introduction of fibres, the composites become less ductile as the molecular rearrangement does not have time to take place. The notching effect of the fibres is also important in which considerable stress concentration is induced in the matrix at the fibre end and matrix flow is constrained by adjacent fibres. Takahashi and Choi [77] who studied on the failure mechanisms showed that under loading of tensile stress, the cracks start at the fibre ends and propagate along the fibre-matrix interface or cross through the matrix and finally the failure takes place.

Curtis *et al.* [73] reported that fibre ends significantly concentrate the stress in the adjacent matrix which increases the stress magnifications. The effect of these stress concentrations can be relieved only by matrix flow, interface debonding or matrix fracture. This shows that, the failure is closely related to the number of fibre ends compared to other parameters such as fibre volume fraction, fibre length and fibre diameter which are used to determine the number of fibre ends. Nonetheless, the relationship between composite failure strain and fibre volume fraction, mean fibre length as well as fibre diameter has been studied by other authors [26, 56, 69, 72, 78].

Most of the theoretical considerations concerning the strength of short fibre reinforced thermoplastics use the assumption that the composite fracture is mainly characterised by either fibre breakage or fibre pull-out combined with matrix fracture. Clean pull-out fibres on the fracture surfaces implied weak fibre-matrix interfaces while profound matrix-adhered fibres denoted strong fibre-matrix interfaces. When the fibre-matrix interfaces are strong, the composites are tougher than unreinforced polymer matrices

[69, 79]. Figure 4.173(a) represents the observation of the fracture surfaces of glass fibre composite which shows that the matrix adhered on the fibre surface as an evidence for the strong fibre-matrix adhesion. Similar observation has been reported by Fu *et al.* [69] in their studies on fibre-matrix adhesion and interphases.

PA 6,6 is hygroscopic materials i.e., moisture sensitive which can absorb moisture up to 2.5 wt% when exposed to humid environments. The amount of moisture absorption at a specific period of time depends on the diffusion coefficients of the individual components of the composites. The kinetic of the diffusion process depends on the temperature and relative moisture absorption. The higher the relative moisture absorption the greater is the absorption rate. This takes place through a diffusion process, in which water molecules are transported from areas of high concentration to areas of lower moisture concentration. This diffusion process can be described by Fick's law [80]. The moisture absorption is usually limited to the matrix while little moisture is absorbed by fibres. Water can also diffuse rapidly into composites along the fibre-matrix interface.

Figures 4.66 and 4.67 suggest that an increase in moisture absorption reduce the composites strength. As presented in Table 4.21, the tensile strength for unreinforced PA 6,6 in dry, 50% RH and wet conditions are found to be 53 MPa, 50 MPa and 35 MPa respectively. 6.2% and 32% decrements reported as samples are conditioned in 50% RH and wet environment respectively compared to SGD0 counterpart.

Tensile strengths of composites for 4% V_f in dry, 50% RH and wet conditions are found to be 89 MPa, 60 MPa and 43 MPa respectively. 32% and 51% decrements in tensile strengths are reported in 50% RH and wet samples respectively compared to SGD4 counterpart. For 8%, 13% and 16% V_f the same trend are observed. The decrements in tensile strengths in 50% RH condition are 39%, 22% and 34% respectively compared to

SGD8, SGD13 and SGD16 counterparts. Furthermore, in wet condition, the tensile strengths reduce about 51%, 52% and 47% compared to SGD8, SGD13 and SGD16 counterparts. This could be due to the fact that the amount of water causes the fibre to swell which alters its dimensions and in turn changes the size, shape, stiffness, and permeability of yarns and fabrics. The moisture acts as a plasticiser that reduces the entanglement and bonding between molecules therefore increases their volume and mobility [81]. Huang and Sun [82] reported that the tensile strength decreased gradually when glass/polyester composites were immersed in water. The reaction between water molecules and the matrix deteriorates the interphase hence weakens the composites. They believed that water can cause matrix swelling and the softening of the matrix material results in decrease of composite strength.

The tensile moduli of these composites decrease dramatically compared to the tensile strength. As can be seen in Table 4.21, the tensile modulus for unreinforced PA 6,6 in dry, 50% RH and wet conditions are found to be 0.71 GPa, 0.62 GPa and 0.38 GPa respectively. The tensile moduli decrease 14% and 47% compared to SGD0 counterpart for 50% RH and wet condition, respectively. The same trend was observed for 4%, 8%, 13% and 16% V_f in both conditions. In 50% RH condition, the decrements in tensile modulus are 62%, 48%, 42% and 26% compared to SGD4, SGD8, SGD13 and SGD16 counterparts. When the samples are conditioned in wet environment, the tensile moduli decrease about 77%, 67%, 58% and 40% compared to SGD4, SGD8, SGD13 and SGD16, respectively.

At higher fibre content, it can clearly be seen that stress transfer capability between fibre and matrix interface gets reduced as the moisture content increases. Mohd Ishak *et al.* [59] have suggested that the absorbed moisture has significantly changed the fracture mode from being brittle to ductile fracture which results in reduction of the tensile strength and tensile modulus. In other studies, Thomason [83] reported that the reduction in tensile strength and tensile modulus is due to the existence of water which lowers the stress transfer capability of the fibre-matrix interface.

The reductions in modulus when the samples are conditioned in wet environment are due to the region created by the presence of water where larger surfaces of fibre and matrix participate in the adhesion process. Abdel-Magid *et al.* [84] speculated that the larger contact area may allow the load to transfer to the fibre more effectively thus maintaining the strength. However, the effect of plasticisation is much greater hence reduces the strength and modulus.

In addition, the reduction in tensile modulus is expected to be due to poor adhesion between matrix resin and fibre interphase when immersed in water. This can be seen in Figure 4.176 whereby the fibre surface is smooth without adhered resin particles, illustrating the reduction of the effectiveness of the interfacial bond. It also showed the matrix softens and debonds at the fibre-matrix interphase. A similar phenomenon has been reported by Huang and Sun [82] for the glass/polyester composite when the sample were immersed in water.

The histograms for fracture strain with function of conditions are illustrated in Figure 4.68 and the values of fracture strain are presented in Table 4.21. The fracture strain for unreinforced PA 6,6 in dry, 50% RH and wet conditions are found to be 231%, 349% and 149% respectively. The fracture strain increases 51% when the samples are conditioned at 50% RH and the fracture strain is found to decrease about 36% when the samples are exposed to wet condition compared to SGD0 counterpart. The reinforced composites in this range of study also show increment of fracture strain in 50% RH and decrease in fracture strain when samples are immersed in water for wet condition. No clear trend is observed as the moisture absorption increase because the tests are carried out without using the strain gauge due to the limitation of the machine.

Generally, the absorbed moisture in polymer increases the ductility of the composites. The increase in failure strain upon exposure of the samples to a wet environment can be attributed to the plasticisation of nylon 6,6 caused by moisture absorption. This is due to the lubrication effect of moisture, which allows the polymer chains to slip past each other. This is because hydrogen bonding between moisture and matrix is formed in the composites and the dipole–dipole interaction between matrix and fibres become less effective [85]. When a fibre breaks, the load it supports is transferred to surrounding fibres by matrix resin. As the relative moisture absorbed is increasing, adhesion between matrix resin and fibre becomes poorer; hence the matrix can no longer effectively distribute the applied load over an appreciable length of the adjacent fibres. Therefore, less occurrence of fibre breakage and thus, fracture strain of composite is increased.

4.3.1.2 Carbon fibre composites

The tensile properties of injection moulded carbon fibre composites are given in Table 4.22 and Figures 4.69 to 4.70. The composite was found to break in brittle manner. The tensile strengths for these composites in dry, 50% RH and wet conditions are 228 MPa, 188 MPa and 118 MPa, respectively. The tensile strength is drastically decreased as the moisture absorption increases. In 50% RH and wet conditions, 17% and 93% decrements respectively in tensile strengths are observed compared to SCD33 counterpart.

The tensile moduli also show the same trend as the moisture absorption increases. The tensile moduli for these composites in dry, 50% RH and wet conditions are 7.0 GPa, 5.0 GPa and 4.5 GPa, respectively. The decrements are 28% when the samples are conditioned at 50% RH and are found to decrease about 37% when the samples are exposed to wet condition compared to SCD33 counterpart.

As expected, the fracture strain is found to increase significantly as the moisture absorption increases. The fracture strains for this composite in dry, 50% RH and wet conditions are 5.5%, 6.1% and 7.2%, respectively. 11% and 24% increments in fracture strain are observed in 50% RH and wet conditions compared to SCD33 counterpart, respectively.

The reduction in tensile strength and tensile modulus and increase in fracture strain are strongly caused by existence of water. Commonly, it is assumed that water diffuses into the amorphous regions of the polymer where hydrolysis should occur at a rate which depends upon the crystallinity and the initial content of end groups [86]. Crystallinity of nylon breaks up intermolecular hydrogen bonding and allows the hydrocarbon chain segments to pack better. Consequently, the absorbed water lowers tensile modulus and strength.

4.3.2 General comparison between single and hybrid fibre composites

To establish whether hybridisation produces a positive or negative effect compared to the properties of single fibre composites, the predicted values of composite strength are calculated using simple rule of mixture (ROM) equation. A direct comparison of data is very difficult due to the variety of fibres used. Most of the properties of hybrid composites obey the rule of mixture but in some cases the measured value is higher than the ROM value.

ROM is a linear equation relating a property of the hybrid to the corresponding properties of the individual single fibre composites. Parameter such as volume fraction of the constituent materials is normally used.

The general equation that has been used is:-

$$P_h = P_c C_c + P_g C_g \qquad (4.3)$$

where, P_h is the predicted property of hybrid composite. P_c and P_g are the experimentally determined properties of single carbon fibre composite and single glass fibre composite respectively. C_c and C_g are the compositions of single carbon fibre and glass fibre of the total fibre content. C_c and C_g are given by equations 4.4 and 4.5 respectively:-

$$C_c = \frac{V_f^c}{V_f^t} \qquad (4.4)$$

$$C_g = \frac{V_f^g}{V_f^t} \qquad (4.5)$$

$$V_f^t = V_f^c + V_f^g \qquad (4.6)$$

$$V_f^t + V_{m,} = 1 \qquad (4.7)$$

where, $V_f{}^c$, $V_f{}^g$, $V_f{}^t$ are respectively the volume fractions of carbon fibre, glass fibre and total fibre in the hybrid composites.

Values of tensile strength, tensile modulus and fracture strain and calculated values using ROM equation are given in Tables 4.23 to 4.25. The plots of the experimental and predicted tensile strength and fracture strain against the glass/carbon fibre composite ratio at different conditions are given in Figures 4.72 to 4.74 and Figures 4.78 to 4.80.

The predicted values of tensile strength and fracture strain are higher than experimental values. Without considering other factors, it can be deduced that, hybridisation shows a negative effect on the tensile strength and fracture strain. The opposite behaviour of the composites has been observed in tensile modulus. It can be seen that the experimental values are higher than the predicted values showing a positive effect on tensile modulus at all conditions (Figures 4.75 to 4.77). Hashemi [87] has also reported the positive hybrid effect on tensile modulus while studying the effect of temperatures on tensile strength and modulus of ABS polymer filled with short glass fibre and glass bead particles.

4.4 Flexural properties

4.4.1 Single fibre composites

4.4.1.1 Glass fibre composites

Flexural testing provides information about materials bending behaviour. In flexural tests, most of the specimens break in ductile manner. The flexural properties versus fibre volume fraction results are shown in Figures 4.81 to 4.83 and the values are reported in Table 4.26. Both flexural strength and flexural modulus are increased for the composites with higher fibre volume fraction.

The flexural strength for unreinforced PA 6,6 in dry, 50% RH and wet conditions are found to be 140 MPa, 110 MPa and 82 MPa respectively. As the amount of fibre loading increases, 22-94%, 34-128% and 28-130% increments in tensile strengths at dry, 50% RH and wet condition are observed compared to unreinforced PA 6,6 counterparts. Furthermore, the flexural modulus for unreinforced PA 6,6 in dry, 50% RH and wet conditions are found to be 2.2 GPa, 1.1 GPa and 0.5 GPa respectively. As the fibre volume fraction increases, 16-186%, 7-311% and 35-269% increments in tensile modulus at dry, 50% RH and wet condition are observed compared to unreinforced PA 6,6 respectively.

The increase in fibre content leads to an increase in flexural strength and flexural modulus which indicates that the glass fibre reinforced PA 6,6 results in better flexural strength and flexural modulus than the pure matrix. This observation can be explained by the contribution of the glass fibre as a brittle and tough material. The loss of ductility is confirmed by increasing glass fibre content. In general, an increase in fibre content results in an increase in stiffness and toughen the composites. Thomason [88] in his work on long glass fibre reinforced PA 6,6 also reported the increment in composite strength as the fibre content increased.

The flexural displacement as a function of fibre volume fraction is shown in Figure 4.83. As presented in Table 4.26 the flexural displacement for unreinforced PA 6,6 in dry, 50% RH and wet conditions are found to be 0.09 mm/mm, 0.13 mm/mm and 0.2 mm/mm respectively. As the fibre volume fraction increases, 26-46%, 8-45% and 55-63% decrements in fracture strains at dry, 50% RH and wet condition are observed compared to unreinforced PA 6,6 counterparts.

Table 4.26 represents the results of the flexural properties and the histograms of the flexural properties versus conditions for different volume fractions are shown in Figures 4.84 to 4.86. As seen in the table, the flexural strength for unreinforced PA 6,6 in dry, 50% RH and wet conditions are found to be 140 MPa, 110 MPa and 82 MPa respectively. The flexural strengths are decrease by 21% and 42% compared to SGD0 counterpart for 50% RH and wet condition, respectively. The same trend is observed for 16% V_f for both conditions. In 50% RH and wet conditions the decrement of flexural strength are 7.7% and 31% compared to SGD16 respectively. However, samples with

4%, 8% and 13% V_f do not follow the same trend as unreinforced PA 6,6 and 16% V_f. They show the increment of flexural strengths in 50% RH condition and decrease in flexural strengths when samples are immersed in water for wet condition. The awkward trend obtained is probably due to optimisation of the interfaces between fibre and matrix which leads to the higher value of flexural strength in 50% RH compared to dry condition.

Generally, the flexural strength drops abruptly as the moisture content increases. This may be caused by the formation of weak hydrogen bonding between the water molecules and PA 6,6 matrix. The water molecules diffuse through the material seeking out any charged areas and forcing polymer chains apart along the way. This is the reason why some parts of polyamide swell after being exposed to moisture. The separation of the polymer chains reduces the polar attraction between chains and increases chain mobility resulting in decrease mechanical properties.

Flexural modulus for 50% RH and wet samples are found to decrease gradually compared to dry samples. As shown in Table 4.26, the flexural modulus for unreinforced PA 6,6 in dry, 50% RH and wet conditions are found to be 2.2 GPa, 1.1 GPa and 0.5 GPa respectively. The flexural moduli are decreased by 49% and 77% compared to SGD0 counterpart for 50% RH and wet condition, respectively. The same trend is observed for 4%, 8%, 13% and 16% V_f in both conditions. In 50% RH condition, the decrements in flexural modulus are 53%, 30%, 33% and 27% compared to SGD4, SGD8, SGD13 and SGD16 counterparts. When the samples are conditioned in wet environment, the flexural modulus are decrease about 69%, 61%, 57% and 50% compared to SGD4, SGD8, SGD13 and SGD16 respectively.

Flexural displacement for 50% RH and wet samples has increase dramatically compared to dry samples. As presented in Table 4.26, the flexural displacement for unreinforced PA 6,6 in dry, 50% RH and wet conditions are found to be 0.09 mm/mm, 0.134 mm/mm and 0.2 mm/mm respectively. The flexural displacements are increased by 43% and 113% compared to SGD0 counterpart for 50% RH and wet condition, respectively. The same trend is observed for 4%, 8%, 13% and 16% V_f in both conditions. In 50% RH condition, the increments in flexural displacement are 76%, 48%, 73% and 45% compared to SGD4, SGD8, SGD13 and SGD16, respectively. When the samples are conditioned in wet environment, the flexural displacement are increased about 30%, 24%, 23% and 47% compared to SGD4, SGD8, SGD13 and SGD16 counterparts.

The composites tend to be ductile when expose to humid environment as the fibre volume fraction decreases. Huang and Sun [82] believed that when the wet samples are being bended, the water molecules would enhance the integrity of the samples as a result of the greater bending resistance. In addition, all the voids will be filled with water and acts as a plasticiser to favour the bending strength. Other authors, Dhakal *et al.* [89] and Stamboulis *et al.* [90] have been reported that water molecules act as a plasticiser agent in composite material which can lead to the increase of the maximum strain for the composites after water absorption.

However, the flexural displacements show the highest value at 50% RH compared to wet condition in reinforced composites samples. This might be due to the existence of microvoids which allow the moisture to fill space and enhance the bending behaviour as reported by Huang and Sun [82].

4.4.1.2 Carbon fibre composites

The results of flexural properties are reported in Table 4.27 where the histograms of the composite at different conditions are shown in Figures 4.87 to 4.89. The flexural strength for carbon fibre composites in dry, 50% RH and wet conditions are found to be 432 MPa, 363 MPa and 255 MPa respectively. The flexural strengths are decreased by 16% and 41% compared to SCD33 counterpart for 50% RH and wet condition, respectively.

Moreover, the flexural modulus is found to decrease as the moisture absorption increases. In dry, 50% RH and wet conditions flexural moduli are 27.1 GPa, 14.0 GPa and 9.3 GPa respectively. 49% and 66% decrements in flexural moduli are observed when samples are conditioned at 50% RH and wet environment compared to SCD33 counterpart.

The flexural displacement for carbon fibre composites in dry, 50% RH and wet conditions are found to be 0.02 mm/mm, 0.04 mm/mm and 0.04 mm/mm respectively. The flexural displacement increases 100% in 50% RH and wet conditions compared to SCD33. Similar flexural displacement values are observed in both conditions. This observation is in agreement with other mechanical properties studies by Huang and Sun [82] on glass/polyester composites which revealed that the bending behaviour of the samples was improved after water immersion. When the composites have been immersed in water, water molecules would be attracted by the hydrophilic groups in the matrix and carbon fibre.

4.4.2 General comparison between single and hybrid fibre composites

Flexural strength, flexural modulus and flexural displacement of glass/carbon hybrid fibre composites together with the calculated values are tabulated in Tables 4.28 to 4.30. The plots of flexural strength, flexural modulus and flexural displacement against composition of carbon fibre composites are illustrated in Figures 4.90 to 4.98.

In general, the experimental values of flexural strength, flexural modulus and flexural displacement of hybrid composites are lower than the predicted values according to the Equation 4.3 in all tested conditions. From these values, without considering other factors, it can be deduced that, hybridisation produces a negative effect on flexural properties. Similar observations are reported by Park and Jang [91] while studying the effects of hybridisation of aramid /polyethylene composites. The authors noted that the increase in aramid fibre content reduces aramid/PE contacts and thermal stresses which contribute to the improvement of flexural strength. The residual stress which is near to the interface between the different fibres could prevent load transfer between fibres and matrix. Hence, when the load is applied to the hybrid composite, the interphase region between two different fibres becomes weak and the composite fails initially in this region which results in the negative hybrid effect.

4.5 Impact properties

4.5.1 Single fibre composites

4.5.1.1 Glass fibre composites

Table 4.31 shows the impact test results of unreinforced PA 6,6 and reinforced composites at dry, 50% RH and wet conditions and the plots of these values against notch to depth ratios (a/D) of various fibre volume fractions are given in Figures 4.100 to 4.105. All specimens are subjected at dry, 50% RH and wet condition. Generally, in dry condition fracture energy and peak load are increased with an increase in fibre volume fraction. Samples which contain higher fibre loading (V_f 16%) give the highest value of fracture energy and peak load compared to unreinforced PA 6,6. Gullu *et al.*

[92] has also reported that, the impact energy was high in 30% weight fraction due to increasing in fibre fractures as the fibre content increased.

However, at low V_f such as 4% and 8% fracture energy and peak load are increased with an increase in fibre volume fraction but their fracture energy are slightly lower than unreinforced PA 6,6. Unreinforced PA 6,6 has failure or argument (due to absence of glass fibre) to react towards high speed loading in impact testing. Hence it gives higher value of fracture energy compared to reinforced composites. In contrast, when the specimens were subjected to 50% RH and wet conditions, the effect of fibre loading on impact properties are slightly change due to water sorption effect.

In impact tests, the energy of high speed loading restricts the ductility of matrix. Glass fibres strengthened the composite specimens, hence more energy is required to break the samples.

Figures 4.105 to 4.114 show the effect of absorbed moisture on all specimens for Charpy impact test and the values of fracture energy, W, peak load,P and notch to depth ratio, a/D are presented in Tables 4.32 and 4.33. It can be seen that the fracture energy and peak load are increased with increase in absorbed water for all composite at different fibre volume fractions. As discussed earlier in section 4.4.1.1, water acts as plasticiser which has a similar effect to the water in breaking down hydrogen bonding in the amorphous region and increases in ductility and flexibility of the composites. Thereafter, it affects the maximum load bearing ability [58, 93].

The plots of fracture energy values against fibre volume fraction, V_f at various notch to depth ratios, a/D are given in Figures 4.115 to 4.120. All specimens are subjected at dry, 50% RH and wet condition. With regard to the effect of increasing notch to depth ratios, a/D the fracture energy and peak load are decreased for all composites in dry, 50% RH and wet condition. This is due to the nature of PA 6,6 which is notch sensitive and the increasing of a/D reduces the fracture area. In general, notches are stress concentrators that reduce the ability of a material to sustain a load and promote brittle, rather than ductile failure [94]. Notched impact energy can be thought of as a measure of crack propagation. The crack propagation will occur along the path of least resistance, requiring as little energy as possible. In the case of composites, the crack may propagate at the matrix and fibre interface [95].

For different a/D specimens, the fracture energy and peak load have smaller values at lower V_f, compared to unreinforced PA 6,6 whereas at higher V_f, greater fracture energy and peak load are observed. This is expected as low fibre loading gives notching effect while at high fibre loading gives reinforcement effect which toughened the composites. However, this trend occurs only for dry samples since there is plasticising effect on 50% RH and wet samples due to existence of water.

The relationship between fracture energy (W), the critical strain energy release rate (G_c), and the specimen geometry parameter (BDΦ) is given by [57] :

$$W = G_c BD\Phi \qquad (4.8)$$

where B and D are the width and depth of the specimen, respectively. A correction factor, Φ is given by:

$$\Phi = \frac{1}{2}\left(\frac{a}{D}\right) + \frac{1}{18\Pi}\left(\frac{S}{a}\right) \qquad (4.9)$$

where a and S are notch depth (or crack length) and span of the specimens, respectively. A plot of W against BDΦ should produce a straight line, where its slope is equal to the critical strain energy release rate, G_c of the materials.

The average variations in value of fracture energy with BDΦ for these composites are given in Figures 4.121 to 4.123. G_c values as taken from the slope of a straight line of these plots are given in Tables 4.34 to 4.36. G_c is also known as fracture energy by Thomason and Vlug [96] and Bijsterbosch and Gaymans [97].

It was noted that, G_c value is increased as the fibre volume fraction increases for the composites at the same condition. Nonetheless, the G_c values for SGD4 and SGD8 are lower than unreinforced PA 6,6 in dry condition. The increment of G_c with V_f is expected to be due to the existence of high fibre content, which further has more tendencies to fail with fibre pull-out mechanism (higher fracture energy) [57].

This observation is in agreement with Thomason and Vlug [96] who studied the influence of the fibre length and concentration on the notched Charpy impact energy of random in plane glass fibre reinforced polypropylene (PP) composites. Bijsterbosch and Gaymans [97] also revealed that the notched Charpy impact energy increases with increase of fibre volume fraction in their study on injection moulded short glass fibre reinforced polyamide 6 composites.

The histogram of G_c values against fibre volume fraction at different conditions are given in Figure 4.125. In general, if the value for unreinforced PA 6,6 is ignored, G_c values are found to increase with an increase in fibre volume fraction. Hassan *et al.* [16] have reported the increment in G_c and K_c values with increase in fibre volume fraction which is expected as more fibres will be controlling the fracture behaviour with increase in fibre loading.

Tables 4.34 to 4.36 show the G_c values and the histogram of G_c values against conditions are given in Figure 4.126. It can be seen clearly that the G_c is increased with increase of moisture content. Samples at the same V_f give the highest value when tested in wet condition compared to 50% RH and dry condition. This observation indicates that absorption of moisture in the composites material has improved the impact strength of the material. In general, water absorption is concentrated on the amorphous regions of the polymer where it has the effect of plasticising the material by interrupting the polymer hydrogen bonding, making it more flexible and increases the impact strength [58]. It differs notably from the behaviour of PET composites [98] where a slight decrease in impact strength can be observed during ageing which may related to the chemical degradation of the polyester matrix.

The average values variations of σY with $1/\sqrt{a}$ of the composites are shown in Figures 4.126 to 4.128. K_c values as taken from the slope of a straight line of these plots are reported in Tables 4.34 to 4.36. K_c provides a measure of tolerance of the material to presence of discontinuities and the plots of K_c values against conditions at different fibre volume fractions are given in Figure 4.129. With regard to the increase in fibre volume fraction, K_c behaves in the same way as G_c i.e. its values are increased.

Tables 4.34 to 4.36 show the K_c values and the histogram of K_c values against fibre volume fraction at different conditions are given in Figure 4.130. Similar behaviour as in G_c when moisture content increases has been observed in 4% V_f and 8% V_f. However, K_c values for unreinforced PA 6,6, 13% V_f and 16% V_f do not show any good trend as the absorbed moisture increases.

4.5.1.2 Carbon fibre composites

The average values of fracture energy, W, peak load, P and notch to depth ratios, a/D for the injection moulded carbon fibre composites are given in Table 4.37. All specimens are subjected to dry, 50% RH and wet condition. The Figures 4.131 and 4.132 show the effect of absorbed moisture on the composite for Charpy impact test. It can be seen that the fracture energy and peak load are increased with increase in absorbed moisture. Similar trend is observed as in glass fibre composite.

The histograms of fracture energy values against conditions at various notch to depth ratios, a/D are given in Figures 4.133 and 4.134. With regards to the effect of increasing notch to depth ratio, a/D the fracture energy and peak load are decreased for the composites in dry, 50% RH and wet condition. Bader and Ellis [99] investigated the impact behaviour of carbon fibre epoxy composite materials using the Charpy impact test of V-notched specimens. They found that there is a critical length-to-depth ratio which marks a change of fracture mode from a relatively brittle transverse cracking to a tougher mode involving extensive delamination.

The variation of average values of fracture energy with BDΦ for these composites is given in Figure 4.135. G_c values as taken from the slope of a straight line of these plots are given in Table 4.38.

The histograms of G_c values against condition are given in Figure 4.136. It can be seen that the G_c is increased when the samples were exposed to humid environment, the same case as in glass fibre composites.

Furthermore, the average values variation of σY with 1/√a of the composites is shown in Figure 4.138. K_c values as taken from the slope of a straight line of these plots are tabulated in Table 4.38. The histograms of K_c values against fibre volume fraction at different conditions are given in Figure 4.138. From the observation, K_c shows the opposite trend from G_c with increase in moisture absorption.

4.5.2 General comparison between single and hybrid fibre composites

The average values of fracture energy, W, peak load, P and notch to depth ratios, a/D for the injection moulded glass/carbon hybrid fibre composites are given in Tables 4.39 to 4.41. The plots of fracture energy against notch to depth ratio subjected to various fibre ratios are illustrated in Figures 4.139 to 4.144. Generally, it can be seen that the fracture energy is increased with increase in glass fibre content.

The toughness of glass fibres is greater than carbon fibres, thus the energy dissipated in the glass fibre is higher. The failure of glass fibre composite is normally accompanied by large scale pull-out of the fibres, though the strain energy stored is relatively small. By contrast, a carbon fibre reinforced thermoplastic has less impact strength than a glass fibre composite where there is hardly any pull-out of fibres in the carbon fibre composite but the strain energy content is higher due to high stiffness of the composite. Thus, this indicates an interesting possibility that both fibres in combination may yield a material with higher fracture toughness.

Fu *et al.* [65], reported that the impact energy increases with an increase of the glass fibre volume fraction and a corresponding decrease of carbon fibre volume fraction which suggest that the addition of glass fibre is more efficient in enhancing the work of fracture of PP than the carbon.

Figures 4.145 to 4.154 show the plots of fracture energy against notch to depth ratio subjected to different conditions and the values of fracture energy, W, peak load, P and notch to depth ratio, a/D are presented in Tables 4.39 to 4.41. It can be seen that the fracture energy and peak load are increased with increase in water absorption for all composites at different fibre ratios. This is due to plasticising effect caused by existence of water in the composites system as discussed earlier.

The plots of fracture energy values against the composition of carbon fibre composites at various notch to depth ratios, a/D are illustrated in Figures 4.154 to 4.160. All specimens are subjected to dry, 50% RH and wet condition. With regard to the effect of increasing notch to depth ratio, a/D the fracture energy and peak load are decreased for all composites in dry, 50% RH and wet condition.

The variation of average values of fracture energy with BDΦ for these composites is given in Figures 4.161 to 4.163. G_c values as taken from the slope of a straight line of these plots are given in Tables 4.42 to 4.44. The plot of G_c values against condition at different compositions of carbon fibre composites is given in Figure 4.164.

As can be seen from the table and figures, the values of G_c decrease with increase in carbon fibre content at 50% RH and wet conditions. In contrast, at dry condition samples SG50/C50 show the highest G_c value compared to SG25/C75 and SG75/C25 hybrid composites.

Thorat and Subhash [100] in their work observed that the hybrids laminates containing less of glass and more of carbon fibres behaves as carbon fibre composite and with reversed composition behaved as glass fibre composite. The authors quoted that this is due to the improvement in the fracture toughness probably attributed to a constructive interaction of a distinctly different nature of failure process in glass fibre composite and carbon fibre composites. In glass fibre composite, a large amount of energy is absorbed in the process of fibre pull-out which is somewhat similar to plastic deformation in ductile materials. In carbon fibre composites, it is predominantly a brittle fracture and most of the energy is linear.

The plot of G_c values against composition of carbon fibre composite at different conditions is given in Figure 4.165. This plot shows the effect of moisture on G_c values of hybrid composites. The G_c values are increased when the moisture exists in the composite system, as similar in the case of single glass and carbon fibre composites.

Moreover, the average values variation of σY with $1/\surd a$ of the composites is shown in Figures 4.166 to 4.168. K_c values as taken from the slope of a straight line of these plots are reported in Tables 4.42 to 4.44. The histogram of K_c values against condition at different composition of carbon fibre composites is given in Figure 4.169. As can be seen, as the carbon fibre content is increased, the values of K_c do not show any trend in all conditions. The histogram of K_c values against composition of carbon fibre composite at different conditions is given in Figure 4.170. The K_c values in wet condition give the lowest values compared to dry and 50% RH condition. No clear trend has been observed as the moisture content increases.

4.6 Fractographic studies

4.6.1 Effect of moisture on tensile fracture surfaces

Figures 4.173 to 4.176 show the micrographs of broken sections of tensile test specimens of glass and carbon fibre composites for dry and wet specimens. From Figure 4.173, no matrix deformation is observed which explains the extreme brittle behaviour of glass fibre composite during tensile test for dry as moulded specimens.

After water immersion (Figure 4.174), it can be seen that the surface of some fibres is smooth and the matrix swells which illustrates the physical damage of the interphase and debonding between fibre and matrix. Moreover, the moisture has inhibited the matrix deformation of the composite. Thus, it is expected that during the failure process, the resistance to crack propagation will be reduced and the samples break in brittle manner as discussed above.

In addition, the figures also show a clear fibre pull-out on both types of composites in dry and wet condition fracture surfaces. This means that shear stresses along the fibre-matrix interface are not high enough to cause the fibre failure after matrix fracture. This phenomenon has been reported by several authors [59,101] who revealed that the failure mechanism is dominated by fibre pull-out. However, it can be seen in Figure 4.175 that there is a good fibre-matrix bonding at carbon fibre surfaces. When the composites were immersed in water (Figure 4.176), the stronger fibre-matrix bonding is observed compared to dry composites. This is expected to be due to the carbon fibre properties itself where the active carbon fibre surface can strongly attract polar molecules of the polymer matrix which tightens the matrix and fibres together [102].

4.6.2 Effect of moisture on impact fracture surfaces

Figures 4.177 to 4.180 show SEM micrographs of the impact fracture specimens for glass and carbon fibre composites at dry and wet condition. The SEM micrographs of impact specimens can provide information concerning the failure mechanisms of the composites. Damage mechanisms such as fibre pull-out can be clearly seen in Figure 4.177. Fibre pull-out is a common feature for the fibre reinforced composites tested at low and high speeds. Fibre pull-out and fibre matrix debonding are known as the primary energy absorbing mechanisms that yield higher fracture energy in fibre reinforced composites [103]. The similar failure mechanism has been observed by Hassan *et al.* [57]

in their study on short and long glass fibre reinforced PA 6,6. The authors noted that the fibre pull-out mechanism will contribute more fracture energy to the composites.

Debonding of the fibres can also be seen in the micrographs. Figure 4.178 show that the glass fibre surface is relatively clean and smooth which indicates the debonding between the glass fibre and matrix. Tjong *et al.* [104] observed that the glass fibre debonding and pull-out are responsible for energy absorbing mechanisms in the PA6,6/SEBS-g-MA hybrids during impact loading. Fibre debonding and pull-out are resulted from a weak fibre-matrix interfacial debonding and from a reduction in fibre length.

Moreover, in the case of carbon fibre, (Figures 4.179 to 4.180) it can be observed that there is a good fibre matrix bonding in the composites. Solomon *et al.* [103] also reported similar finding and the authors quoted that this might be due to the better nucleating capabilities of carbon fibre compared to the glass fibre. It results in better bonding properties between carbon fibre and PA 6,6 matrix when this property coupled with the rough carbon fibre surface. After immersion in water, the similar matrix deformation as in tensile mode can be observed for both type of fibres.