A comparison of as-fatigue and re-consolidation residual properties for notched quasi-isotropic $[0/45/90/-45]_{2S}$ and cross-ply $[0/90]_{4S}$ AS4/PEEK composite laminates

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Abstract

A comparison of the as-fatigued and re-consolidated properties have been made between notched quasi-isotropic $[0/45/90/-45]_{2S}$ and cross-ply $[0/90]_{4S}$ AS4/PEEK laminates. For the former, the $\pm 45^\circ$ plies tend to constrain longitudinal damage development so that damage growth primarily occurred in the transverse direction, causing more widespread damage. This led to prominent mechanical properties degradation, shorter fatigue lives and lower residual strengths. For cross-ply laminates, quick and extensive longitudinal crack tangential to the hole and the corresponding 90° fiber shear off brought about effective stress concentration alleviation. This discouraged further damage development. Hence, their fatigue lives exceeded one million cycles even at high cyclic stress levels and their residual strengths were significantly higher than their virgin strength. On the other hand, the re-consolidation process removed most of the defects that alleviated the stress concentration and thus decreased the strengths. Detailed study of the residual strength changes and damage development history revealed that the residual as-fatigued and re-consolidated strengths were governed by the competition between local structural decay and its resulting stress concentration alleviation.

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1. Introduction

Fatigue of notched plates showed that damage invariably occurs in the region near the hole edge and this causes a localized deterioration of mechanical properties. It also leads to a redistribution of stress ahead of the notch to give a more uniform net section stress. The former effect may directly result in a degradation in residual strength over the virgin laminate, namely the ‘wear-out’ effect. However, the latter phenomenon may improve strength and fatigue life, and is called the ‘wear-in’ effect. Whether the as-fatigued residual strength increase or decrease is dependent on the combined effect of these two mechanisms. Notch strengthening under cyclic loading caused by the combined effect of wear-in and wear-out mechanisms has been studied by Stinchcomb and co-workers [1–6]. A re-consolidation method had been employed on the thermoplastic resin matrix composites to isolate the influences of these two

mechanisms in notched $[0/90]_{4S}$ AS4/PEEK laminate [7]. For such lay-up sequence, quick extensive development of longitudinal cracks tangential to the hole edge and the corresponding 90° fibers shear off effectively brought forth the wear-in mechanism and raised the residual strength. This enabled the fatigue life to extend to beyond one million cycles even when fatigued at a loading 90% the strength of virgin specimen. Residual strengths after re-consolidation are invariably lower than their as-fatigue counterparts, presumably because most damages have been healed, and the accompanying beneficial stress-smoothing wear-in mechanism removed.

The mechanical properties of composites are strongly dependent on the stacking sequence of the laminate. During the past two decades, the as-fatigue properties for different lay-up sequence have been widely studied [8–10] but the properties following re-consolidation have yet received limited attention. In this work, the as-fatigue and re-consolidated properties of two different lay-up of notched laminates, quasi-isotropic $[0/45/90/-45]_{2S}$ (laminate Q) and cross-ply $[0/90]_{4S}$ (laminate C) will be studied, and

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the results will be used to evaluate the role played by wear-in and wear-out mechanisms.

2. Experimental works

2.1. Specimen preparations

Unidirectional AS4/PEEK prepregs were laid up in the stacking sequences Q $[(0/45/90/-45)_{2S}]$ and C $[(0/90)_{4S}]$. The prepreg has nominally 61% by volume of graphite fiber, and is 125 μm thick. The stacks were fabricated at 390 °C under 10 kg/cm² pressure in a diaphragm type forming mold according to the conditions recommended by ICI. Cooling of the formed laminates was two staged. Firstly it was inside the hot forming mold at a rate of about 3 °C/min to 270 °C, and secondly at a rate of about 50 °C/min to room temperature by transferring to a water cooled mold. Pressure over the diaphragm was maintained except during the transferring process. The resulting nominal plate thickness is 2.1 mm. All specimens, with width 25.4 mm and length 210 mm, were cut from the plate using a water mist-cooled diamond-coated wheel. A center hole of 6.3 mm diameter was drilled in each specimen. To alleviate artificial damage induced during the drilling of center notch [11], specimens were sandwiched between two sacrificial plates. Water-cooled Ti-coated high-speed steel drills were used for hole machining. Woven glass/epoxy end-tabs were glued to both ends of the specimen using cyanoacrylic adhesive for gripping. Free specimen length between end tabs was 120 mm. Furthermore, two anvils 25.4 mm apart straddling the hole were adhered to the specimen for extensometer sitting.

2.2. Experimental techniques

Static and fatigue tests were performed on an MTS 810 servo-hydraulic machine at room temperature. Static tensile tests were done according to ASTM D3039M. For fatigue tests, tension–tension loading with a stress ratio ($\sigma_{\min}/\sigma_{\max}$) of 0.1 and frequency of 10 Hz were adopted. An extensometer was used to monitor the stiffness for static tests, and the fatigue modulus [12] for fatigue tests. For each type of laminate, the applied stress level is defined as the maximum applied cyclic stress, $\sigma_{\max}$, normalized by the ultimate tensile strength, σ_0 , of the corresponding virgin specimen.

A re-consolidation mold that allows individual tested specimens to be re-consolidated has been proven to be able to remove all of the damage inside the specimen except fiber breakage [7,13]. During the re-consolidation process, all conditions except the pressure are the same as that for primary forming. The re-consolidation pressure used was 5 kg/cm². Microscopic and ultrasonic examination showed that all damages except 0° fiber breakage have been removed. This will isolate the influences of the wear-in

and wear-out mechanisms to a certain degree because most of the wear-in effect will be removed as damage heals yet the major wear-out effect, i.e. 0° fiber breakage, remains. A resin rich region is formed at each of the fracture sites. Moreover, small voids were found in larger resin rich sites. Re-consolidated virgin specimens revealed no change in strength and matrix crystallinity.

In order to evaluate the strain distribution in the remaining ligament of some specimens, a series of five strain gages were placed ahead of the notch edge. In addition, a gage was placed sufficiently upstream from the notch to measure the nominal strain of the specimen. Ultrasonic C-scan with a 15 MHz probe was carried out to examine the damage generated by fabrication and machining, and monitor the damage generated by loading on each specimen. Moreover, a number of the fatigued specimens were examined by tetrabromoethane penetrant-X-ray radiography to complement the C-scan results. In addition, some of the fatigued specimens were cut and mounted to examine internal damages using optical microscope.

3. Results and discussions

3.1. Static tests

The notched tensile strengths of laminates Q and C are 404 and 474 MPa with a coefficient of deviation of 3.21 and 2.25%, respectively. For laminate Q, 0° fiber breakage and 45° slip lines near the hole edges started to be visible to the naked eye at about 97% of its ultimate tensile strength. However, no visible damage was evident before final failure for laminate C. For each type of laminates, one sample was examined using X-rays while loaded incrementally to 80, 90, 95 and 97% of its ultimate tensile strength. A little triangular shaped damage zone was observed ahead of notch edge for laminate Q only until 95% of its ultimate tensile strength, but no damage was evident in laminate C even up to 97% of its ultimate tensile strength. From the above observations, it suggests that initiation, propagation and catastrophic failure in laminate C almost occur concurrently while these processes are progressive in laminate Q.

3.2. Fatigue tests

Fatigue tests were carried out using 65–85% stress level at 5% steps. Tests were terminated at one million cycles if no failure occurred. For these cases, three more specimens were tested to ascertain that fatigue life at the corresponding stress level was unambiguously over one million cycles. If failure occurred, at least eight specimens were used for that stress level to evaluate the scatter in fatigue lives. Fig. 1 shows the S–N curves for both types of laminates. It is worth noting that the lifetimes of laminate C invariably exceeded one million cycles even at 90% cyclic stress level. However, laminate Q only lasted for ~1050 cycles at 85%

Fig. 1. S–N curves for the notched quasi-isotropic $[0/45/90/-45]_{2S}$ and cross-ply $[0/90]_{4S}$ AS4/PEEK laminates.

stress level. As the stress level decreased to 70%, the fatigue life of Q was over one million cycles.

For each type of laminates, at least one specimen tested at each stress level was examined periodically using ultrasonic C-scan during fatigue. Typical results for both laminates are shown in Fig. 2(a) and (b). For laminate Q, two semilunar shaped damages invariably initiated from the notch edges after ~ 500 cycles. They then propagated quickly in the transverse direction, i.e. perpendicular to the loading direction, even at the lowest stress level of 65%. After reaching a certain cycles, propagation was then mainly limited to the longitudinal direction. For the cases of 75, 80 and 85% stress levels, longitudinal cracks and fiber breakage were observable by the naked eye near the hole before final fracture. For the same amount of fatigue modulus loss, the damage region increases with the cyclic stress level. For laminate C, four pronounced longitudinal cracks tangential to the hole were evident. Moreover, such damage can be observed by the naked eye after ~ 500 cycles even at the lowest stress level of 65%. The crack length increased quickly in the first 100 000 cycles. Above 500 000 cycles, the crack length remained nearly constant.

Fig. 2. Typical C-scan examination results of (a) quasi-isotropic $[0/45/90/-45]_{2S}$ and (b) cross-ply $[0/90]_{4S}$ laminates at 75% stress level.

For the same amount of cycling, the length and width of the longitudinal cracks increase with the cyclic stress level. The longest longitudinal crack length observed was 18.01 mm. These lengths were reproducible to within ± 1.5 mm when the testing conditions were the same. Two specimens with shortened end tabs to increase the free specimen length to 160 mm have been cycled with 85% stress level for 1 000 000 cycles. The longitudinal crack lengths obtained were the same as those obtained with normal specimen length within normal experimental scatter. Arrest of the longitudinal cracks may be attributed to the alleviation of stress concentration effect as damages accumulate and as the cracks grow away from the notch. Detailed discussion on these will be given in the latter sections. The differences in damage modes in the two types of laminates may explain their extremely difference in fatigue lifetimes.

The fatigue modulus, defined as the maximum instantaneous stress divided by the corresponding strain, was monitored during cyclic loading. Fig. 3(a) and (b) shows the development of normalized fatigue modulus versus the normalized life for laminates Q and C. The fatigue modulus $F(N)$ at any cycle N is normalized by the initial fatigue modulus F_0 . The cycle number N is normalized by the respective fatigue life N_f . When N_f exceeds one million cycles, N is normalized with one million. Each curve shown in Fig. 3 is the average from all the specimens tested at

Fig. 3. Normalized fatigue modulus versus normalized fatigue life for different stress levels: (a) quasi-isotropic $[0/45/90/-45]_{2S}$ and (b) cross-ply $[0/90]_{4S}$ laminates.

the corresponding stress level. In Fig. 3(a), the fatigue modulus of laminate Q at 65% stress level falls rapidly first and then decreases very gently subsequently. At 1 000 000 cycles, the fatigue modulus has only dropped by 12%. The same trend is found for the case of 70% stress level. However, for the cases of 75, 80, and 85% stress level, moduli were continuously degrading at fairly steady rate until close to final failure, whereby it decreased steeply. In Fig. 3(b), all degradation curves for laminate C show the same trend as the cases of 65 and 70% stress levels of laminate Q. Fatigue modulus is primarily controlled by the integrity of reinforcing fibers. Initial steep drop in this modulus suggests the occurrence of fiber breakage, which presumably occurred near the hole edge. This in turn has the effect of alleviating stress concentration and postponing further fatigue damage. If this alleviation is larger enough, further fatigue modulus decrease will be very gentle and fatigue life will be long. Typical examples are the 65 and 70% stress level cases in laminate Q, and all the cases in laminate C as are evident in Fig. 3(a) and (b). If the beneficial stress concentration alleviation effect is overwhelming the stiffness wear-out effect due to fiber breakage, steady continuous fatigue modulus decay will result. The latter is exemplified by the 80 and 85% stress level cases in laminate Q. It is interesting to note that the 75% stress level case of laminate Q represents a transition from the former to the latter phenomenon.

3.3. Residual strengths

Three stress levels, 65, 75 and 85%, were adopted for as-fatigued and re-consolidated residual strengths examinations. The category with long fatigue lives (all laminate C and 65% stress level laminate Q) was examined on a cycle basis, with as-fatigued and re-consolidated residual strength tests after 100 000, 500 000 and 1 000 000 cycles. For the category with short fatigue lives (laminate Q at 75 and 85% stress level), the strengths were examined after 10 and 20% fatigue modulus losses. The normalized as-fatigued and re-consolidated residual strengths of laminates Q and C for various damage states at different stress levels are compared in Figs. 4 and 5, respectively. Each data point is the average from at least three specimens.

The as-fatigued residual strength of the 65% stress level laminate Q first increased by 3% up to 500 000 cycles and decreased to near the virgin strength at one million cycles (Fig. 4(a)). The residual strengths for laminate Q tested at the higher stress levels are invariably lower than their virgin strengths. Furthermore, the residual strength decreases as the stress level and/or modulus loss increase. Contrarily, all as-fatigued residual strengths of laminate C are significantly above their virgin strengths (Fig. 4(b)). Moreover, this trend is more marked if cycled at a higher stress level or for a larger amount of cycles. To make a comparison between both types of laminates, the above trends are consistent with the previous fatigue modulus examinations. For the case

Fig. 4. As-fatigued residual strengths of (a) quasi-isotropic [0/45/90/-45]_{2S} and (b) cross-ply [0/90]_{4S} laminates.

that no final failure occurred at one million cycles, their residual strengths were invariably above or equal to their virgin strengths. This means that the wear-in effect outweighed or balanced the wear-out effect. For the other cases that failed at much shorter fatigue lives, the wear-out effect overwhelmed and residual strengths were lower than virgin strength.

After re-consolidation, the residual strengths of both types of laminates are invariably lower than their as-fatigued counterparts (Fig. 5(a) and (b)). This phenomenon is presumably resulted from the wear-in effect being removed during re-consolidation, while the major wear-out effect (i.e. fiber breakage) still remains and governs the strength. For the same stress level, the re-consolidated strengths of laminates Q decreased monotonically with the previous fatigue cycles or modulus loss. For laminate C, the trends of re-consolidated strength versus previous elapsed cycles are more complicated (Fig. 5(b)). These seemingly different trends can be understood as a result of the competitive wear-in and wear-out effects due to remnant and newly induced defects. Detailed argument and evidence are given in Ref. [7]. In the next sections, optical microscopy observations and strain distribution ahead of the notch tip on both types of laminates

Fig. 5. Re-consolidated residual strengths of (a) quasi-isotropic $[0/45/90/-45]_{2S}$ and (b) cross-ply $[0/90]_{4S}$ laminates.

will be used to explain the re-consolidated mechanical behaviors and to explain the difference between laminates Q and C.

3.4. Damage observations

For each type of laminate, a loaded virgin specimen, some as-fatigued specimens, and some monotonically re-loaded re-consolidated specimens were taken for optical microscopy observation. These specimens were sectioned into two halves. Mounting of the samples was so made to expose the front, edge and top views on the left or right half. Progressive polishing down the mounted samples revealed the extent of damage at different layers and different positions ahead of the notch. For laminate Q, the as-fatigued specimens examined included two that had been fatigued for 500 000 and 1 000 000 cycles (65% stress level) and two with 20% modulus loss at 75 and 85% stress levels, respectively. For laminate C, samples were selected from each stress level and had been fatigued for 500 000 and 1 000 000 cycles. The re-consolidated samples examined were taken from every condition tested in the last section. Moreover, these re-consolidated samples had been loaded

monotonically once to 80% (laminate Q) or 90% (laminate C) of their respective virgin strengths.

3.4.1. Laminate Q

No damages were found up to 80% stress level in loaded virgin specimen. This is consistent with the previously mentioned X-ray examinations where damages were only observed at 95% of its ultimate tensile strength. For the as-fatigued specimens, 0° fiber breakage and longitudinal cracks invariably occurred in the region near the hole (see F and L in Fig. 6(a)), and were more serious and concentrated with increasing cyclic stress level. For $\pm 45^\circ$ layers, matrix cracking and fiber breakage were evident, but the amount of

Fig. 6. Optical microscopic observations of notched $[0/45/90/-45]_{2S}$ laminate: (a) 0° , (b) 45° plies and (c) edge view (75% stress level for 20% fatigue modulus loss).

fiber breakage is not so extensively as in the 0° layer (see C and F in Fig. 6(b)). Matrix cracking was also found in the 90° layers. In addition, extensive delamination (see D in Fig. 6(c)) was observed in the edge view. Delamination was more serious with higher stress levels. Progressive polishing down the mounted samples revealed no damage at ~ 2.5 mm away from the notch in the edge views for all three stress levels.

For the re-consolidated specimens, fractured fibers with resin rich regions and voids at the fiber fracture sites were observed. When loading was re-applied once, in addition to the defects just named, matrix cracking (*M*), fiber/matrix interfacial debonding (*I*) and delamination (*D*) were found to have developed near the clustered fiber fracture sites. Typical results are shown in Fig. 7(a) and (b). It is obvious that the newly induced damages for the case of 85% stress level are more extensive and serious than those for the 65% case. Such damages were never found in as-re-consolidated specimen, and were probably caused by the stress concentration effect due to the remnant defects, such as fiber breakage and voids [7]. In other words, by loading once, a whole range of defects are re-generated. This in turn will bring back most of the wear-in effect and so the correspondingly re-consolidated residual strengths do not deviate much from the as-fatigued strengths.

Fig. 7. Edge view of the re-consolidated quasi-isotropic $[0/45/90/-45]_{2S}$ specimens after static loading to 80% of its virgin strength. Previous fatigue histories: (a) 65% stress level for 500 000 cycles and (b) 85% stress level to 20% fatigue modulus loss.

3.4.2. Laminate C

For laminate C, no damages were found up to 90% stress level in loaded virgin specimen. This is also consistent with the previously mentioned X-ray examination where no damages were found even up to 97% of its ultimate tensile strength. For the as-fatigued specimens, longitudinal crack (*L*) tangential to the hole edge invariably occurred in the 0° layers and surrounded by delamination (Fig. 8(a)). As the stress level increases, the width of the longitudinal cracks also increases. This agrees with the indirect observation using ultrasonic C-scan. Besides, 0° fiber breakage (*F*) is evident at all three stress levels, being more serious and concentrated for the 85% stress level (Fig. 8(a) and (c)). Extensive delamination is observed at the 85% stress level (see D in Fig. 8(c)), however, it is very limited in the other two stress levels. In the 90° layers, fiber breakage and matrix cracking are both observed (see F and C in Fig. 8(b)). Fiber fracture occurred in the positions corresponding to the crack in the 0° layers. No damage was found away from the longitudinal crack. It is therefore reasoned that such damage may be caused by the longitudinal crack, which creates significant stiffness difference on its two sides and induces matrix cracking and fibers in the 90° layers shear off. Opening of the matrix cracks increases with the applied stress level but these cracks are very short and occur sparsely except at the highest stress level of 85%. Progressive polishing down the mounted samples revealed that damage ceased to be observed at 0.5, 0.5 and 0.9 mm from notch edge in the edge view for 65, 75 and 85% stress levels, respectively.

Laminate C exhibits extensive longitudinal cracks (see L in Fig. 8(a)) that grow in preference to transverse damage development probably because the comparatively smaller interlaminar toughness versus the transverse fracture toughness [14,15]. Such extensive longitudinal cracks will shear off the 90° fiber (see F in Fig. 8(b)) and lead to a fast and effective reduction of the stress concentration. This may be the reason for fatigue life of laminate C exceeding one million cycles even when fatigued at 85% stress level. Laminate Q also shows the tendency of such longitudinal crack (see L in Fig. 6(a)). However, their lengthwise development was restrained by the 45° fibers. This prevented quick alleviation of stress concentration. New longitudinal crack and other damages were thus induced ahead of the notch, making the damage zone spread in the transverse direction. After reaching a certain transverse damage zone size, stress has effectively been redistributed. Further damage development along the longitudinal direction become easier than the transverse direction due to the comparatively smaller interlaminar toughness versus the transverse fracture toughness. Thus, near the end of fatigue life when the transverse damage zone is large, tendency of damage extending along the longitudinal direction is evident (see 19 000 and 27 000 cycles in Fig. 2(a)).

By monotonically loading the re-consolidated specimens to 90% of the virgin strength, new damage was not evident

Fig. 8. Optical microscopy observations for notched $[0/90]_{4S}$ laminate in (a) 0° ply, front view, (b) 90° ply, edge view and (c) edge view. (85% stress level for 500 000 cycles).

in all the 65 and 75% stress levels cases and in the 85% case with a previous fatigue cycle of 100 000. For the cases of 85% stress level with an elapsed 500 000 and 1 000 000 cycles, many matrix cracks developed in the 90° layers.

Fig. 9. Edge view of the re-consolidated cross-ply $[0/90]_{4S}$ specimens after static loading to 90% of its virgin strength. Previous fatigue histories: (a) 500 000 and (b) 1 000 000 cycles at 85% stress level.

Fiber/matrix interfacial debonding in the 0° layers and a little delamination were also observed (see I and D in Fig. 9(a) and (b)). Such damages were more extensive in the 1 000 000 cycles case but were never found in as-re-consolidated specimens. Without the re-creation of damage, the beneficial wear-in effect is seriously checked, leading to a re-consolidated strength being much lower than the as-fatigued strength. This phenomenon is markedly different from that in laminate Q where the newly induced damages invariably occurred in all cases due to the existence of the more extensive remnant defects.

3.5. Notch strain distributions

Five strain gage arrays with gage length 1 mm were attached to the virgin, as-fatigued and re-consolidated specimens of both types of laminates to examine the strain distribution ahead of notch tip (Fig. 10). Specimens in the latter two categories had been fatigued under 65% (for 1 000 000 cycles) and 85% (for 1 000 000 cycles and 20% modulus loss for laminates Q and C, respectively) stress levels. For laminate Q, all specimens were loaded in 10% step to 80% of its virgin strength, i.e. $0.8\sigma_0^Q$ with the superscript denoting laminate Q. Laminate C was loaded likewise to $0.9\sigma_0^C$.

Fig. 10. A series of strain gages to measure the strain distribution ahead of the notch edge.

3.5.1. Laminate Q

For the case of 65% stress level of laminate Q, the strain distributions at $0.8\sigma_0^Q$ are shown in Fig. 11. It is obvious that the strain distribution in the virgin specimen is fairly close to the anisotropic elasticity analysis [16]. It changed linearly with the applied load even up to $0.8\sigma_0^Q$. For the as-fatigued specimen, the strain distribution also changed linearly with the applied load up to $0.8\sigma_0^Q$. At $0.8\sigma_0^Q$, gage I recorded a decrease of $3312 \mu\epsilon$ compared with the virgin specimen, however, gage II recorded an increase of $3992 \mu\epsilon$ and the other gages recorded slight increases. Previously, C-scan examinations showed that the damaged region was within the positions of gages I and II. The local stiffness of the damaged region would be degraded. In the face of a degraded local stiffness at gage I position, a decreased local strain suggests that the stress in this region has been decreased. In other words, the stress concentration effect has been alleviated and stress has been redistributed to the other part of the net section. Damage observation showed that at

Fig. 11. Strain distributions ahead of notch for virgin, as-fatigued and re-consolidated quasi-isotropic $[0/45/90/-45]_{2S}$ specimens (65% stress level and fatigued for 1 000 000 cycles).

gage II position most of the damage was matrix cracking and delamination. Local stiffness decrease due to defects in the minor load carrier is expected to be small. Such large increase in strain at gage II position has to be attributed to increasing local stress. Since positions corresponding to gage III–V are way off the damaged region, no significant local stiffness change is expected there. Strain increased in this region therefore confirms that the alleviated stress concentration has been redistributed here. This tendency of leveling off of stress across the net section is expected to postpone specimen failure to a higher load, giving rise to the so-called wear-in phenomenon. As a result, although extensive and serious fatigue damages were observed that should significantly wear-out the specimen strength, the residual strength still preserved the virgin level (see the 65% case in Fig. 4(a)). In other words, the wear-in and wear-out effects almost balanced each other in this case.

For the re-consolidated specimens, the strain distribution started to change nonlinearly when the applied stress exceeded $\sim 0.4\sigma_0^Q$. At the same time, some ‘quack’ sound was continually heard all the way up to $0.8\sigma_0^Q$. It suggests that damages may be reformed from the remnant defects during this period. Strain distribution in the re-consolidated specimen revealed a slight strain increase at gage I while a large increase of $4531 \mu\epsilon$ at gage II when compared with the as-fatigued counterpart. The strains at gages III–V were slightly lower than those of the as-fatigued specimen. It implies that the newly induced damages were not alleviating the stress concentration to the same extent as that of the as-fatigued specimen. This corroborates with the observation that the re-consolidated residual strength being lower than its as-fatigued counterpart (see the 65% case in Fig. 5(a)).

In the following, we compare the strain of the as-fatigued specimen cycled at 85% stress level against that of the virgin one at the same locations and loading. At $0.1\sigma_0^Q$ gage I recorded an increase of $6652 \mu\epsilon$. On reaching $0.2\sigma_0^Q$, gage I was broken. At $0.8\sigma_0^Q$, gage II recorded an increase of $2017 \mu\epsilon$. Gages III–V recorded increases of strain over those of the virgin specimen. The latter increases are more marked than that in the 65% stress level case. It therefore suggests that alleviation of stress concentration is more effective in the 85% case. The increases in strain at gage I was attributable to serious local stiffness degradation. However, even in the face of this beneficial stress-redistributed, the as-fatigued residual strength was lower than its virgin strength (Fig. 5(a)). This may be explained by the fact that fracturing of the major loading carrier (0° fibers) was more extensive and serious than those of the 65% case. For the re-consolidated specimen, the stress–strain relation became nonlinear when the applied load exceeded $\sim 0.3\sigma_0^Q$. During re-loading, gage I recorded continuous increase with the applied load because the local stiffness degradation due to the serious remnant defects, especially the fiber breakage. The strains at the other gage location varied in the same trend as the case of 65% stress level. At $0.8\sigma_0^Q$, the strains at gages III–V were slightly

lower than those of the as-fatigued specimen. Incidentally, the re-consolidated residual strength was lower than its as-fatigued counterpart (see the 85% case in Fig. 4(a)).

3.5.2. Laminate C

Again the strain distribution in the virgin specimen is also fairly close to the anisotropic elasticity analysis [16] and changed linearly with the applied load even up to $0.9\sigma_0^C$. For the as-fatigued and re-consolidated specimens, the strain distributions also changed linearly with the applied load up to $0.9\sigma_0^C$. Fig. 12 shows the strain distributions measured at $0.9\sigma_0^C$ for specimens which had undergone one million cycles. Since the stiffness of laminate C is much higher than that of laminate Q, stress changes at gages III–V positions will be reflected as much smaller strain variations. In addition, the damage zone in the transverse direction of laminate C is limited to near the hole edge. Hence, only the strain recorded by gages I and II were shown in Fig. 12 for clarity. For the as-fatigued specimen (65% stress level), gage I recorded a decrease of $820 \mu\epsilon$ compared with the virgin specimen, however, gages II–V recorded slight increases. Previously C-scan observations showed that the damaged region was well within the gage I position and the local stiffness will be degraded. Decrease in strain thus implies that the stress concentration effect has been alleviated and stress has been redistributed to the positions of gages II–V. As with laminate Q, the tendency of stress leveling off is certainly expected to postpone specimen failure to a higher load. This is consistent with the previous as-fatigued residual strength examinations. Moreover, because the damages were limited in the transverse direction, the residual strengths were invariably above its virgin strength (Fig. 3(b)), namely the wear-in effect invariably outweighed the wear-out effect. The same trend but more prominent results were found in the case of 85% stress level. At $0.9\sigma_0^C$, gages I and II recorded a decrease of 1840 and $50 \mu\epsilon$ compared with those of the virgin specimen, while gages III–V recorded increases. This

implies more effective stress redistribution and corroborates with the previous observations that the as-fatigued strength increases more with cycling at a higher stress level.

After re-consolidation, the strain distributions of both cases changed linearly with the applied load even up to $0.9\sigma_0^C$. For the case of 85% stress level, although the newly induced damages were observed, they were comparatively limited in extent (Figs. 7(b) and 9) and not enough to influence the strain distribution. On reaching $0.9\sigma_0^C$, gage I in general recorded a rise in strain over the as-fatigued specimen, because the beneficial wear-in effect was removed. This is accompanied with a reduction in the residual strengths (Fig. 5(b)). For the case of 65% stress level, strain indicated by gage I has nearly increased back to the virgin level. However, strain at gage I in the other cases is lower than that for the virgin specimen. This implies that certain alleviation of stress concentration caused by the remnant defects was still existent. In other words, although no significant newly induced damage occurred, the remnant defects, i.e. fractured fiber, still provided beneficial wear-in effect to a certain degree. Moreover, because the remnant defects caused by the previous cyclic loading were limited to the transverse direction, the scatter bands of the re-consolidated residual strengths largely overlapped with its virgin strength for most of cases (Fig. 5(b)).

4. Conclusions

Virgin tensile and fatigue behavior, as well as the residual properties after fatigue and re-consolidation, have been examined on notched quasi-isotropic $[0/45/90/-45]_{25}$ and cross-ply $[0/90]_{45}$ AS4/PEEK laminates. It was found that

1. No damage was observed in laminate Q when loaded to 95% of its ultimate tensile strength. Thereafter, damage extended progressively across the net-section toward final failure. For laminate C, no damage was observed even up to 97% of ultimate tensile strength, initiation and the catastrophic completion of failure almost occurred concurrently.
2. Two typical modes of modulus degradation have been identified in fatigue testing. One mode showed an initial sharp drop followed by a very slow decrease. This was invariably associated with long fatigue lives exceeding a million cycles. The other mode has a gently but steadily decreasing modulus that showed accelerated rate of drop towards the end of life. The latter mode was associated with much shorter fatigue lives.
3. For the as-fatigue tests, quick and extensive longitudinal crack and the corresponding 90° fiber shear off in laminate C effectively caused the alleviation of stress concentration and thus refrained damage from spreading in the transverse direction. Thus, the residual strengths were significantly above their virgin strength. For

Fig. 12. Strain distributions ahead of notch in the virgin, as-fatigued and re-consolidated cross-ply $[0/90]_{45}$ specimens fatigued for 1 000 000 cycles.

laminate Q, alleviation of stress concentration has also occurred. However, the $\pm 45^\circ$ fibers tend to constrain damages development in the longitudinal direction. Damages mainly spread in the transverse direction, thus breaking much more amounts of major load carriers. As a result, their residual strengths were always lower than its virgin strength.

4. For both laminates, the re-consolidated residual strengths were invariably lower than their as-fatigued counterparts because most of the beneficial effect of the wear-in mechanism has been removed. The residual strength after re-consolidation was governed by the competitive wear-in and wear-out effects resulting both for remnant defects and newly induced damages. Since new damages, i.e. wear-in effect, invariably occurred in laminate Q due to the existence of the more remnant defects, the correspondingly re-consolidated residual strengths do not deviate much from the as-fatigued strengths.

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