Evaluation of detailed transverse crack propagation toward thickness direction for CF/PEEK quasi-isotropic laminates under fatigue loading

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Abstract
Carbon fiber (CF) reinforced polyether ether ketone (PEEK) laminates are applied to the members of transportation structures exposed to severe conditions, because carbon fibers and PEEK have excellent heat-resistant and chemical stability. Therefore, it is important to ensure the long-term reliability of CF/PEEK laminates and evaluate the propagation of transverse cracks, which are generally the initial damage incurred by multidirectional laminates under fatigue loading. This study evaluated the detailed mechanism of transverse crack propagation toward the thickness direction in CF/PEEK quasi-isotropic laminates under fatigue loading. The transverse crack growth was observed using the replica method, and the energy release rate associated with the transverse crack propagation toward the thickness direction was calculated using a three-dimensional virtual crack closure-integral method. Thus, it was found that the transverse crack propagated in a stable manner until it passed through the 90° layer toward the thickness direction. Additionally, it was found that the crack growth rate increased in the range wherein the crack length was short, and maintained an approximately constant growth rate in the middle stage of propagation. According to Paris’ law, when the crack depth was relatively shallow, the experimental results are approximately consistent with the analytical evaluation results for the energy release rate associated with the transverse crack propagation toward the thickness direction. Finally, it is also concluded that the high ductility of the PEEK resin and good fiber/matrix interfacial property of the CF/PEEK laminates are important influencing factors for the stable and characteristic crack propagation observed in the experiment.

Keywords: Transverse crack, Crack propagation, Fatigue, Energy release rate, Paris’ law, CF/PEEK

1. Introduction

Carbon fiber reinforced thermoplastic (CFRTP) composites have high mass productivity, moldability, and good energy absorption performance compared with carbon fiber reinforced thermoset (CFRTS) composites. Thus, they can be applied to the members of transportation structures. CFRTP laminates with PEEK as the matrix resin (CF/PEEK) can be applied to the members of transportation structures exposed to severe conditions, because CF/PEEK laminates have good heat-resistant and chemical stability amongst CFRTP laminates. To apply the CF/PEEK composites to these parts, the evaluation of their long-term reliability is indispensible. Thus, the fatigue characteristics of CF/PEEK laminates have been investigated.

Previous studies have reported that the fracture mechanism of CF/PEEK 0° unidirectional (UD) laminates under...
fatigue loading is different to that of CFRTS (CF/Epoxy) laminates, owing to the low CF/PEEK interfacial strength (Gamstedt and Talreja, 1999). Moreover, it has been reported that the fatigue life of the CF/PEEK 0° UD laminates is inferior to that of CF/Epoxy because the stress-sharing ratio at the fiber breaking points remains high owing to the ductility of the PEEK resin, which causes more fiber breaks under fatigue loading (Zhang and Hartwig, 2002). As mentioned above, even in 0° UD laminates, for which the fiber characteristics are considered to be relatively dominant, the fiber/matrix interface properties and matrix resin properties affect the fatigue life. When applying CF/PEEK laminates to structural members, multidirectional laminates, such as quasi-isotropic (QI) laminates, are typically used. For multidirectional laminates, a transverse crack occurring at the 90° layer surface is generally the initial damage and leads to more severe damage, such as delamination and ultimately fracture. Several studies have experimentally (Hojo, et al., 1994) and analytically (Naghipour, et al., 2011) evaluated the delamination growth of CF/PEEK laminates under fatigue loading and reported that the delamination growth is restricted owing to the ductility of the PEEK resin, compared with CF/Epoxy laminates under fatigue loading (Hojo, et al., 1994; Aymerich and Found, 2000). With regard to transverse cracks, which are the origin of delamination, it has been reported that their initiation and multiplication in CF/PEEK laminates under fatigue loading is restrained, compared with those of CF/Epoxy laminates (Aymerich and Found, 2000; Song and Otani, 1998; Hosoi, et al., 2018). Notably, the abovementioned studies have only observed the macro internal progression of transverse cracks and considered the toughness of the matrix resin as the cause for the transverse crack initiation and multiplication differences. With regard to the neat PEEK resin, studies focusing on the fatigue crack initiation and propagation mechanism (Shrestha et al., 2016; Sobieraj et al., 2010) and relationship between the fatigue life and PEEK resin properties (Saib et al., 1993) have been conducted. However, the detailed mechanism of transverse crack initiation and propagation, and the influence of the matrix resin properties on the transverse crack initiation and propagation, have not been extensively investigated. These evaluations can provide useful information for developing CF/PEEK laminates with a better fatigue property in terms of transverse crack initiation and propagation restriction.

This study investigated the detailed mechanism of transverse crack propagation toward the thickness direction of CF/PEEK QI laminates under fatigue loading. Experimental and analytical approaches were used to observe the crack propagation and evaluate the energy release rate associated with the propagation of transverse cracks toward the thickness direction.

2. Experimental method
2.1 Specimen

CF/PEEK quasi-isotropic (QI) laminates, for which MR50R is applied as the carbon fiber, were formed using an autoclave. The mold temperature of the CF/PEEK laminates was 653 K and the stress-free temperature was 612 K. The fiber volume fraction of the CF/PEEK was 58.8%, and the specimen’s laminate configuration was [45_2/0_2/-45_2/90_2]. The specimen geometry is shown in Fig. 1. The mechanical properties of the CF/PEEK laminates are listed in Table 1. Emery paper tabs with a length of 50 mm were attached to both ends of each specimen. Each specimens edge surface was polished using emery paper and buffed with diamond powder until the average roughness reached 0.07±0.01 μm.

2.2 Test method and conditions

In this study, the initiation and propagation of transverse cracks in the CF/PEEK QI laminates under tensile fatigue loading was observed using a replica film. The replica method is a technique whereby the surface irregularities of objects are transferred by attaching a replica film whose single-side is softened or dissolved by an organic solvent. Therefore, the image on the replica is horizontally inverted relative to the object’s surface. The replica film consists of cellulose acetate and has a thickness of approximately 0.1 mm. First, replica films were attached to the glass plates to ensure the stability of the films, and acetone as an organic solvent was applied to one side of the films that were not attached to the glass plate. Next, the films that had softened with acetone were attached to the specimen and the replicas were sampled. The middle 10 mm of the specimen’s one edge was defined as the observed area and the replicas were continuously sampled in that area at every 2000 or 4000 cycles (Fig. 2). During the replica sampling, the acetone evaporated into the atmosphere. Subsequently, the transverse crack propagation at the observed area of the specimen’s edge was evaluated by observing the replica films using a digital microscope. The transverse cracks were observed together after the transverse crack had passed through the 90° layer toward the specimen’s thickness direction. By comparing the replicas...
sampled before and after, certain transverse cracks were identified amongst the different replicas. Fatigue tests were conducted with a frequency of $f = 5$ Hz and stress ratio of $R = 0.1$ (tensile-tensile) using a hydraulic fatigue testing machine. The maximum stress of specimen for CF/PEEK QI laminates under cyclic loading, $\sigma_{\text{max}}$, was controlled to 0.6, 0.7, or 0.8 times the stress of specimen, $\sigma_b$, at which the first transverse crack initiated under static loading. The average $\sigma_b$ for the CF/PEEK QI laminates was 411 MPa. Additionally, the transverse cracks were observed using soft X-ray photography to investigate the crack propagation toward the width direction. On the other hand, the fracture surface of 90° UD laminates was observed to evaluate the fiber/matrix interface property under fatigue loading. The stress ratio was $R = 0.1$ and the maximum stress of specimen for CF/PEEK 90° UD laminates, $\sigma_{\text{max}}$, was 0.4 times the static tensile stress of specimen, $\sigma_b$, $\sigma_{\text{max}}/\sigma_b = 0.4$. The average $\sigma_b$ in the CF/PEEK 90° UD laminates was 86.6 MPa.

Table 1. Mechanical properties of CF/PEEK UD laminates.

| Property                        | Symbol | Unit  | Value       |
|---------------------------------|--------|-------|-------------|
| Longitudinal Young’s modulus    | $E_L$  | GPa   | 154         |
| Transverse Young’s modulus      | $E_T$  | GPa   | 9.62        |
| In-plane shear modulus          | $G_{LT}$ | GPa  | 5.60        |
| In-plane Poisson’s ratio        | $\nu_{LT}$ |     | 0.301      |
| Out-of-plane Poisson’s ratio    | $\nu_{TT}$ |     | 0.49*       |
| Longitudinal thermal expansion coefficient | $\alpha_L$ | $[\times10^{-6}/K]$ | 1.05 ($T < 483K$), 6.16 ($T > 483K$) |
| Transverse thermal expansion coefficient | $\alpha_T$ | $[\times10^{-6}/K]$ | 37.0 ($T \leq 411K$), 65.6 ($T > 411K$) |

*Assumed value.

Fig. 1. Geometry of quasi-isotropic (QI) laminate specimen ($[45/0/\pm45/90]_s$).

Fig. 2. Sampling of replica. The edge surface of the CF/PEEK QI specimen was sampled using the replica method below the middle 10 mm of the specimen’s edge. Considering the stability of the replica film, one side of the replica film on which the surface of specimen was not transferred, was attached to the glass plate before sampling. The transverse crack propagation toward the thickness direction was evaluated by observing the transverse cracks on the replica sampled at every 2000 or 4000 cycles.
3. Experimental results

Figure 3(a) to (c) shows a digital microscopy photograph of the replica, wherein the transverse crack on the CF/PEEK QI specimen was transferred at each cycle number. Figure 3(d) shows a digital microscopy photograph of the specimen’s edge after the fatigue test ($N=1.5\times10^5$). Each image in Fig. 3 shows the same specimen location. Figure 3(d) is horizontally inverted for comparison with Fig. 3(c). According to Fig. 3, the transverse crack on the replica propagated toward the thickness direction as the number of cycles increased. The shape of the transverse crack on the replica corresponds well with that on the specimen. Thus, the edge transverse crack propagation toward the thickness direction was evaluated using the replica method under each stress level. However, it is impossible to observe the fiber arrangement in the replica images.

Figure 4(a) to (f) shows the relationship between the number of cycles and the transverse crack length toward the thickness direction for the CF/PEEK QI laminates. Figure 4(a) and (b) show the results obtained at a lower stress level ($\sigma_{\text{max}}/\sigma_i=0.6$). Figure 4(c) and (d) show the results at $\sigma_{\text{max}}/\sigma_i=0.7$, while (e) and (f) are the results obtained at a higher stress level ($\sigma_{\text{max}}/\sigma_i=0.8$). Figure 4(b), (d), and (f) show enlarged views of Fig. 4(a), (c) and (e), respectively. As shown in Fig. 4(a) to (f), for each specimen (each stress level), two or three transverse cracks passed through the 90° layer, and there was no crack initiation on the replica. The crack shown in Fig. 3 corresponds to crack 1 in Fig. 4(c) and (d). Additionally, the value obtained by dividing the transverse crack length toward the thickness direction $a$ by the thickness of the 90° layer $b_0$ is defined as the normalized crack length $a'$. At a lower stress level ($\sigma_{\text{max}}/\sigma_i=0.6$), two transverse cracks coalesced into one during the propagation. Notably, the length of the longer crack was evaluated before the two cracks coalesced.

According to Fig. 4, as a general tendency, the propagation behaviors of these cracks have the following characteristics in common. First, each crack stably propagated toward the thickness direction until the normalized crack length $a'$ was equal to 1 (until it passed through the 90° layer), except for when the cracks coalesced. Secondly, each crack repeatedly progressed and stagnated within the range wherein the crack length was short (approximately $a'<0.4$), and the crack growth rate appeared to increase from a macroscopic view within this range. It is thought that crack repeats propagation and stagnation because crack propagates taking a long way along the fibers. Thirdly, each crack propagated with an approximately constant crack growth rate, regardless of the crack length, within a certain range (approximately $0.4<a'<0.8$, as shown in Fig. 4(b), (d), and (f)). Subsequently, in some cracks, the crack growth rate slightly decreased just before passing through the 90° layer. Therefore, it was confirmed that, at the testing stress level, the transverse cracks exhibited the abovementioned propagation behavior. This behavior was considered along with the analytical results presented in chapter 4. However, at a higher stress level ($\sigma_{\text{max}}/\sigma_i=0.8$), there existed a large crack growth rate dispersion amongst the cracks. Overall, the crack growth rate tended to be lower at a higher stress level ($\sigma_{\text{max}}/\sigma_i=0.8$), compared with that at $\sigma_{\text{max}}/\sigma_i=0.7$. Hence, it is considered that the differences amongst the specimens, such as the dispersion of the

![Fig. 3](image_url)

Fig. 3. (a), (b), and (c) Replica images observed using digital microscopy (d); specimen edge image observed using digital microscopy. The images of the specimen edge were horizontally inverted to align the direction with the replica images. The specimen’s edge surface was transferred almost exactly using the replica method; the transverse crack propagation toward the thickness direction was evaluated using the replica method.
Fig. 4. Relationship between number of cycles and transverse crack length toward thickness direction under each stress level for CF/PEEK laminates; (a) and (b) correspond to a lower stress level ($\sigma_{\text{max}}/\sigma_0=0.6$); (c) and (d) correspond to $\sigma_{\text{max}}/\sigma_0=0.7$; (e) and (f) correspond to a higher stress level ($\sigma_{\text{max}}/\sigma_0=0.8$). The normalized crack length $a'$ represents the value dividing the transverse crack length toward the thickness direction $a$ by the 90° layer thickness $t_90$. The transverse cracks stably propagated until passing through the 90° layer toward the thickness direction. At a lower stress level ($\sigma_{\text{max}}/\sigma_0=0.6$), two transverse cracks coalesced into one during the propagation. Notably, the length of the longer crack was evaluated before the two cracks coalesced. As a general tendency, the transverse cracks repeatedly progressed and stagnated in the range wherein the crack length was short (approximately $a'<0.4$), and propagated with an approximately constant crack growth rate within a certain range (approximately $0.4<a'<0.8$) under the testing stress level and regardless of the crack length. For some cracks, the crack growth rate decreased just before passing through the 90° layer toward the thickness direction.
Fig. 5. Soft X-ray photography of CF/PEEK QI specimen on which transverse cracks were observed using the replica method. The number of cycles was $1.5 \times 10^5$. This confirms that the transverse cracks did not progress into the $90^\circ$ layer toward the width direction for more than approximately 0.3 mm after passing through the $90^\circ$ layer toward the thickness direction and after several thousands of cycles. Delamination was not observed using soft X-ray photography.

fiber distance, affected the behavior of crack propagation toward the thickness direction. As shown in Fig. 3(d), there is the dispersion of fiber distance in $90^\circ$ layer. If the fiber/matrix interface has enough strength, it is thought that the closer the fiber distance is, the more rapid the crack propagates toward the thickness direction due to stress singularity which appears near the fiber/matrix interface. On the other hand, if the fiber/matrix interface strength is weak, it is thought that the closer the fiber distance is, the more slowly the crack propagates toward the thickness direction because crack propagates taking a long way along fibers. The fiber/matrix property of CF/PEEK laminates is considered in chapter 5.

Figure 5 shows a soft X-ray photograph of the CF/PEEK QI specimen at the stress level of $\sigma_{\text{max}}/\sigma_t = 0.7$ after crack 1, crack 2, and crack 3 in Fig. 4(c) passed through the $90^\circ$ layer toward the thickness direction. The specimen shown in Fig. 5 was loaded for $N=1.5 \times 10^5$ cycles, and each transverse crack passed through the thickness direction before $N=1.4 \times 10^5$ cycles had been completed. Figure 5 confirms that the transverse cracks did not progress into the $90^\circ$ layer by more than approximately 0.3 mm toward the width direction after passing through the $90^\circ$ layer toward the thickness direction and after several thousands of cycles had been completed. Therefore, it is concluded that the edge effect influenced the propagation behavior. Additionally, delamination is not observed in Fig. 5 within a range of less than $N=1.5 \times 10^5$ cycles.

4. Analytical evaluation

Based on the results discussed in the previous section, it is concluded that the transversal crack stably propagated toward the thickness direction, and that the transverse crack growth rate toward the thickness direction did not increase when accompanied by the crack growth in every propagation stage for the CF/PEEK QI laminates. Particularly, it can be seen that the transverse crack propagated with an approximately constant growth rate within a certain middle propagation range. This fact suggests that the range of energy release rate $G$ or stress intensity factor $K$ during the fatigue cycle ($\Delta G$ and $\Delta K$) was approximately constant within the range wherein the transverse crack growth rate $d\alpha/dN$ was approximately constant according to Paris’ law (Eq. (1)). To analytically evaluate the transverse crack propagation using Paris’ law, we calculated the energy release rate associated with the transverse crack propagation toward the thickness direction using a type of finite element analysis, namely, the three-dimensional virtual crack closure-integral method (3D VCCM).

$$\frac{da}{dN} = C (\Delta K)^m = C' (\Delta G)^{m'}$$

($C$, $C'$, $m$, and $m'$ are constants)  (1)
The 3D VCCM is an analytical approach proposed by Shivakumar et al. (1988), who expanded the 2D VCCM proposed by Rybicki et al. (1977). The advantage of this method is that the energy release rate associated with the crack propagation can be derived using only the nodal force and displacement calculated for each mode. Although the VCCM for tetrahedral elements has been proposed by various studies (Okada et al., 2007a; 2007b), quadrilateral or hexahedral elements are typically applied to the VCCM. The equation for calculating the energy release rate and reference image are shown below (Eq. (2) and Fig. 6). Equation (2) only applies to linear hexahedral elements. Here, \( G_i \) is the energy release rate when a crack propagates the linear element \( i \); \( F_i \) is the nodal force on the node shared by the linear elements \( i \) and \( i + 1 \); \( v'_i \) is the total displacement of the node adjacent to the node shared by the linear element \( i \) and \( i + 1 \); \( v''_2 \) is the total displacement of the node adjacent to the node shared by the linear element \( i \) and \( i - 1 \); \( \Delta \) and \( w_i \) represent the size of the linear elements \( i \).

\[
G_i = \frac{1}{2\Delta w_i} \left( C_1 F_i v'_i + C_2 F_{i+1} v''_2 \right) \quad \left( C_1 = \frac{w_i}{w_i + w_{i+1}}, \quad C_2 = \frac{w_i}{w_i + w_{i-1}} \right)
\]  

(2)

In this study, the energy release rate was evaluated by 1/2 of the model of the CF/PEEK quasi-isotropic laminates ([±45/0/45/90]s), using the commercial finite element code COMSOL, as shown in Fig. 7. The model consisted of a 45\(^\circ\) layer, 0\(^\circ\) layer, -45\(^\circ\) layer, and 90\(^\circ\) layer, in this order from outside. The dimensions of each layer were set to 20 mm \( \times \) 10 mm \( \times \) 0.15 mm. Linear hexahedral elements were used, the number of elements was 360,000, and the minimum element size was 6.25 \( \mu m \). Figure 7 is magnified toward the thickness direction, owing to the high aspect ratio of each layer. The mechanical properties listed in Table 1 were applied to the model. A rectangular transverse crack was introduced to one side of the 90\(^\circ\) layer edge. The crack depth was set to 0.05 mm, 0.1 mm, and 0.2 mm, according to the experimental results for the crack propagation toward the width direction, as shown in Fig. 5. Additionally, the side surface shape was rectangular, which is compatible with the hexahedral element. The point at the center of the 90\(^\circ\) layer’s surface, where a transverse crack was not introduced, was constrained. Thermal residual stress was applied to the model by reflecting the cooling from stress-free temperature (612K) to room temperature (298K). Based on the thermal expansion coefficients listed in Table 1, the thermal residual stress between the abovementioned temperatures was applied to the model. Using the model, first, the edge effect was evaluated, that is, the stress singularity close to the free edge of the 90\(^\circ\) layer, which affects the transverse crack propagation toward the thickness and width directions. In the evaluation of stress singularity at the edge of the specimen, the transverse crack was not incorporated into the model for the loading condition (\( \sigma_0 \sigma_n = 0.7 \)). Subsequently, the same load (\( \sigma_0 \sigma_n = 0.7 \)) was applied to the model, the mode I energy release rate associated with the transverse crack propagation toward the thickness direction was calculated based on the assumption that the crack depth is constant, and the behavior of the energy release rate toward the thickness direction was investigated for each crack depth. Finally, the transverse crack propagation was experimentally and analytically evaluated using Paris’ law.

![Fig. 6 Reference image of 3D VCCM for hexahedral elements. \( F_i \) is a nodal force on the node shared by the linear elements \( i \) and \( i + 1 \); \( v'_i \) is the total displacement of the node adjacent to the node shared by the linear element \( i \) and \( i + 1 \); \( v''_2 \) is the total displacement of the node adjacent to the node shared by the linear element \( i \) and \( i - 1 \); \( \Delta \) and \( w_i \) represent the size of the linear elements \( i \). The coordinates correspond to that of Fig.7.](image-url)
Fig. 7. CF/PEEK QI laminate model for evaluating energy release rate associated with transverse crack propagation toward thickness direction. A transverse crack with a rectangular side shape was introduced to one side of the 90° layer’s edge. The dimensions of each layer were set to 20 mm × 10 mm × 0.15 mm, and linear hexahedral elements were used in the model. The number of elements was 360,000, and the minimum element size was 6.25 μm. The model is magnified toward the thickness direction based on the high aspect ratio. The depth of the transverse crack was set to 0.05 mm, 0.1 mm, and 0.2 mm based on the result shown in Fig. 5. The thermal residual stress and same load as the experiment (σ/σt=0.7) were applied to the model, and the mode I energy release rate at each crack length was evaluated.

Figure 8 shows the evaluation results for the edge effect, which represents the relationship between the distance from the free edge of the 90° layer and the applied normal stress toward the loading direction in the 90° layer. According to Fig. 8, the applied stress in the 90° layer remained approximately constant when the distance from the edge surface was more than approximately 0.4 mm. In other words, the edge effect was at work only within a range of approximately 0.4 mm from the surface. With regard to the experimental result shown in Fig. 5, it is concluded that the transverse crack propagated into the 90° layer by less than approximately 0.3 mm toward the width direction, just after passing through the 90° layer toward the thickness direction. Therefore, the transverse crack propagated into the 90° layer toward the thickness direction within the region wherein the edge effect first appeared, while the transverse crack propagated mainly toward the width direction after passing through the 90° layer toward the thickness direction.

Figure 9 shows the results of evaluating the energy release rate G associated with the transverse crack propagation toward the thickness direction using 3D VCCM. Figure 9(a), (b) and (c) shows the analytical results at the crack depths of 0.05 mm, 0.1 mm, and 0.2 mm, respectively, and Fig. 9(d) shows a graph combining Fig. 9(a) to (c). During the propagation toward the thickness direction, it was assumed that the crack depth was constant, owing to the edge effect shown in Fig. 8. According to Fig. 9(a) to (c), it is certain that the energy release rate G increased approximately linearly at the early stage of the crack propagation toward the thickness direction for each crack depth. However, when the crack depth was relatively small, it can be seen that the energy release rate G remained approximately constant within a certain crack length range. Additionally, the energy release rate G tended to decrease just before the crack reaches the 90°/−45° interface. As described above, the crack growth rate toward the thickness direction increased when the crack length was relatively short, retained an approximately constant value in the middle stage of propagation, and tended to slightly decrease just before the crack reaches the 90°/−45° interface, with the exception of some cracks. The fluctuation behavior of the energy release rate G corresponds to that of the range of the energy release rate during the fatigue cycle ΔG, because the energy release rate G at any applied stress was calculated by considering the nodal force and displacement, which monotonically increased as applied stress increased. Consequently, according to Paris’ law expressed in Eq. (1),
Based on the experimental results shown in Fig. 5, the transverse crack propagated toward the thickness direction until it passed through the 90° layer within the region where the edge effect appeared, before the crack mainly propagated toward the width direction.

Fig. 8. Relationship between distance from free edge toward width direction and applied stress in 90° layer. The applied stress was approximately constant when the distance from the edge surface was more than approximately 0.4 mm. The edge effect almost does not appear

the experimental result for the crack growth rate was approximately consistent with the analytical result for the energy release rate $G$, when the crack depth was relatively shallow. Additionally, according to Fig. 9(d), the energy release rate $G$ tended to increase as the crack depth increased, because the crack opening constraint by the bottom edge of the crack weakened as the crack depth increased. Thus, according to Paris’ law, it is likely that the crack growth rate toward the thickness direction increased when the crack depth increased. The experimental results, wherein the increase of the crack growth rate is restrained within a certain crack length, confirmed that the crack propagated toward the thickness direction until it passed through the 90° layer without propagating much toward the width direction. In other words, it is concluded that the experimentally observed crack propagation occurred because the crack depth remained relatively shallow.

Next, we comprehensively evaluated the experimental and analytical results using Paris’ diagram. Figure 10 shows Paris’ diagram on the experimental results for the crack growth rate shown in Fig. 4 and analytical results for the range of the energy release rate during the fatigue cycle $\Delta G$ that were calculated using 3D VCCM with the model shown in Fig. 7. The crack growth rate for each crack length was calculated using the secant method, because the replicas were sampled and the crack length was measured ($\Delta N$) at a constant interval. The crack depth was set to 0.05 mm, 0.1 mm, and 0.2 mm and the stress level ($\sigma_{\text{max}}/\sigma_0$) was set to 0.6, 0.7, 0.8 in the model. The broken line represents the approximate line calculated using the least squares method for all plots in each figure. The sharp increase of crack length at a lower stress level ($\sigma_{\text{max}}/\sigma_0=0.6$), when the two cracks coalesced, was excluded. Figure 10 shows that the crack growth rate dispersion was relatively large for each crack depth. Although the plots in Paris’ diagram should align if the transverse crack propagation completely follows Paris’ law, a clear alignment was not observed. One of the reasons for this is that the model for evaluating the range of the energy release rate during the fatigue cycle $\Delta G$ did not consider the actual dispersion of the fiber distribution.

Although the abovementioned results were obtained, the results for Paris’ parameter ($C$ and $m$’ in Eq. (1)) obtained using the least squares method were evaluated for convenience and are listed in Table 2. Figure 11 shows the comparison between the experimental results for the crack propagation shown in Fig. 4 and the results of predicting the crack propagation toward the thickness direction under each stress level. The crack propagation was predicted using Paris’ parameters listed in Table 2, and Eq. (3) derived from the integral of Eq. (1), as follows:

$$N_f - N_i = \int_{a_i}^{a_f} \frac{1}{C' (\Delta G)^m'} \, da \,.$$

(3)
In Eq. (3), subscript $i$ denotes that the crack passed through the 90° layer toward the thickness direction, while subscript $f$ indicates that an initial transverse crack occurred. The number of cycles after which the initial transverse crack occurred at crack depth of $a_i$ is defined as 0.1 times the thickness of the 90° layer $(t_{90})$, which is the minimum length at which most cracks sampled on the replica could be observed. The exact value of $a_i$ is the order of magnitude for the diameter of fibers. According to Fig. 11, the prediction for the number of cycles required for the transverse crack to pass through the 90° layer did not change considerably regardless of the crack depth difference. Additionally, the crack growth rate and number of cycles required for the transverse crack to pass through the 90° layer were reasonably predicted at $\sigma_{\text{max}}/\sigma_i=0.6$ and (c) $\sigma_{\text{max}}/\sigma_i=0.8$. However, there exists a significant difference between the experimental result and the prediction at (b) $\sigma_{\text{max}}/\sigma_i=0.7$. It is considered that the difference between the experimental result and prediction at $\sigma_{\text{max}}/\sigma_i=0.7$ is caused by dispersion in the crack growth rate originating from the difference in the specimens. Therefore, by eliminating the specimen variations as much as possible, we can obtain a useful crack propagation prediction for material design or development. Additionally, it may be possible to compare or evaluate the fatigue property in a transverse crack with the $C'$ and $m'$ obtained using the proposed analysis method for different types of composite materials.

Fig. 9 Behavior of energy release rate $G$ associated with transverse crack propagation toward thickness direction obtained using 3D VCCM: (a), (b) and (c) represent the crack depths of 0.05 mm, 0.1 mm, and 0.2 mm, respectively; (d) graph combining (a) to (c). The energy release rate $G$ increased at the early stage of crack propagation in each crack depth and remained almost constant within a certain crack length range when the crack depth was relatively small. According to Paris’ law expressed in Eq. (1), the overall behavior of the energy release rate tended to be consistent with the experimental crack propagation results shown in Fig. 4. The energy release rate $G$ tended to increase with the crack depth, possibly because the constraint of the crack opening by the bottom edge of the crack weakened as the crack depth increased.
Fig. 10 Paris’ diagram. The crack growth rate toward the thickness direction $\frac{da}{dN}$ was calculated based on the experimental results shown in Fig. 4 using the secant method. The range of the energy release rate $\Delta G$ was calculated by the model shown in Fig. 7 using 3D VCCM; (a), (b) and (c) represent the crack depths of 0.05 mm, 0.1 mm, and 0.2 mm, respectively. The crack growth rate dispersion was relatively large, and a clear alignment, which should exist if the crack propagation completely followed Paris’ law, was not observed. It is thought as the reason that the homogeneous orthotropic model that evaluated the range of the energy release rate $\Delta G$ does not consider the actual dispersion of fiber distribution.

Table 2. Paris’ parameters $C'$ and $m'$ calculated using Paris’ diagram in Fig. 10 and the least squares method.

| Crack depth mm | $C'$   | $m'$  |
|----------------|--------|-------|
| 0.05           | $1.13 \times 10^{-11}$ | 0.809 |
| 0.1            | 0.771  | 0.790 |
| 0.2            | 1.13   | 0.682 |
Discussion

This section discusses the influence of the CF/PEEK laminate properties on the transverse crack propagation mechanism. As shown in Fig. 9, the crack depth greatly affected the energy release rate associated with the transverse crack propagation toward the thickness direction, and also affected the propagation behavior. Thus, it is concluded that the transverse crack in the CF/PEEK laminates propagated in a stable manner, and the crack growth rate stopped increasing at a certain crack length because the crack depth remained relatively shallow. One of the reasons for this is that the crack propagation toward the thickness direction is superior to the crack propagation toward the width direction owing to edge effect. However, as described above, other factors related to the CF/PEEK laminate properties, such as the ductility of the PEEK resin and the fiber/matrix interfacial property of the CF/PEEK laminates, may also have been at work. Figure 12 shows the typical fracture surface of the CF/PEEK 90° UD laminates under fatigue loading. The stress ratio $R$ was 0.1 and the maximum stress $\sigma_{\text{max}}$ was 0.4 times the static tensile stress, $\sigma_t$. The average $\sigma_t$ in the CF/PEEK 90° UD laminates was 86.6 MPa. As shown in Fig. 12, the number of final breaking cycles was $N=2.30\times10^6$. The fiber, matrix, resin, and molding conditions of the 90° UD laminates were the same as those of the QI laminates used to evaluate...
the transverse crack propagation.

As shown in Fig. 12, cohesive failure, rather than fiber/matrix interfacial failure, mainly occurred; therefore, the fiber/matrix interfacial strength was sufficiently high in the CF/PEEK laminates. However, when the fiber/matrix interfacial adhesive property is relatively poor, it is considered that the transverse crack propagates toward the thickness direction as the adjacent interfacial debondings coalesced. In this case, it is thought that the crack depth increases compared with the laminates, which have a good interfacial property, because the interfacial debonding progresses toward the width direction along the fiber/matrix interface. Moreover, the transverse crack propagates into the matrix resin mainly when the interfacial strength is sufficiently high. The transverse crack propagation could have affected the matrix resin ductility when the crack mainly propagated into the matrix resin. Therefore, the ductility of the PEEK resin and the fiber/matrix interfacial property of the CF/PEEK laminates resulted in the abovementioned crack propagation behavior, as confirmed by the experimental results. Furthermore, as can be seen in Fig. 3, the crack propagated in a meandering manner under the effect of the stress distribution between the fibers. These phenomena may have caused the large variation between the experimental and analytical results. It is thought that this discussion is also applicable to composite QI laminates which have 90° layer in the center and edge effect appears.

The crack propagation toward the thickness direction dealt with this study is the earliest stage of damage accumulation under fatigue loading. As cracks propagate and pass through the 90° layer to the thickness direction, stress concentration at the crack tip causes delamination and fiber breakage. These serious damages affect the ultimate fatigue life of the CF/PEEK laminates. By further research on the mechanism of transverse crack initiation and propagation, it is expected that the crack initiation and propagation will be restrained and the ultimate fatigue life of CF/PEEK laminates will be improved.

![Fig. 12: Fracture surface of CF/PEEK 90° UD laminates under fatigue loading (σmax/σb =40%, Nf = 2.3×106). Cohesive failure, rather than fiber/matrix interfacial failure, was mainly observed; therefore, the fiber/matrix interfacial strength was sufficient in the CF/PEEK laminates.](image)

6. Conclusion

This study conducted a detailed experimental and analytical evaluation of transverse crack propagation toward the thickness direction in CF/PEEK QI laminates.

According to the experimental results obtained using the replica method, the transverse crack propagated in a stable manner until it passed through the 90° layer toward the thickness direction under each stress level. Additionally, crack repeatedly propagated and stagnated within the range wherein the crack length was short, and the crack growth rate toward the thickness direction remained approximately constant in the middle stage of propagation as a general tendency. However, the crack growth rate was influenced not only by the stress level, but also by the dispersion amongst the specimens.

According to the analytical results obtained using 3D VCCM, the energy release rate behavior associated with the transverse crack propagation toward the thickness direction was influenced by the crack depth, and the energy release rate behavior was approximately consistent with the experimental crack propagation results when the crack depth was relatively shallow. Paris’ diagram was drawn based on the experimental and analytical results. A clear alignment according to Paris’ law was not confirmed, which reflects the dispersion of the crack growth rate. Additionally, the Paris’ parameters $C'$ and $m'$ were calculated from the Paris’ diagram to predict the crack propagation toward the thickness direction.
direction. Hence, the crack growth rate and number of cycles required for the transverse crack to pass through the 90° layer could be predicted to some extent. However, to improve the prediction accuracy, it is essential to reduce the dispersion amongst the specimens. Moreover, based on the experimental and analytical results, it is concluded that the high ductility of the PEEK resin and fine fiber/matrix interfacial property of the CF/PEEK laminates were important influencing factors for the stable and characteristic crack propagation observed in the experiment.

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