Low-cycle fatigue of the light advanced materials

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Abstract

Low-cycle fatigue (LCF) of 6061-T6 Al alloy plated with nickel, gold and silver as well as thermally-cycled carbon-epoxy laminate was studied at constant strain in a purely bending mode. The tensile and fatigue properties of the alloy coated with multi-layered deposits depend on the thickness of the inner electroless nickel layer that drastically decreases the ductility of the system. Electroplating by the second ductile Ni layer increases the lifetime of the alloy in comparison with that coated only by the electroless Ni layer. Thermal cycling of the carbon fiber reinforced polymer (CFRP) composite at maximum and minimum operating temperatures of 180°C and −195.8°C, respectively, shortens the fatigue life of samples in the high-cycle fatigue range corresponding to the number of cycles $N$ exceeding 20,000. However, in the low-cycle fatigue region, the fatigue life of thermally-cycled laminates was slightly longer than that of the reference one, probably, due to the stress distribution between small cracks, which retards the fatigue failure and prevents fast propagation of the main crack.

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1. Introduction

Coated Al alloys often have a relatively low resistance to cycling load [1-3]. For example, the coating brittleness and the cracks induced during anodizing are among the factors which affect the fatigue strength of hard anodized components [2, 3]. Electroless nickel (EN) is an engineering coating, normally used because of its excellent corrosion and wear resistance. Chemical and physical properties of the deposit vary primarily with phosphorus content and subsequent heat treatment [1, 4]. The results obtained for 7075-T6 [5] and 2618-T61 [6] aluminum alloys coated with an EN deposit show that the coating can give rise to a significant improvement in the fatigue performance of the substrate at medium and low stresses. This improvement was associated with a higher strength of the coating as compared to the substrate and with the development of compressive residual stresses in the coating during the deposition.

All abovementioned literature data demonstrating the fatigue behavior of coated alloys relate to the stress-life method, which works well if only elastic stresses and strains are present. However, most coated components may appear to have nominally cyclic elastic stresses, but various stress concentrators, such as microcracks, notches, welds, etc., present in the component may result in local cyclic plastic deformation. Under these conditions, the local strain as the governing fatigue parameter (the local strain-life method) is much more effective in predicting the fatigue life of a component. In engineering applications, relatively low-frequency strain cycling as a consequence, e.g., of start and stop operations, generates low-cycle fatigue (LCF) failure.

For polymeric composites used in aircraft industry, the material will be inevitably exposed to cyclic thermal loading. Such loading is defined as “thermal fatigue”, which is considered as low-cycle fatigue because thermal fatigue cracks usually start after less than 50,000 cycles [7]. It is known that relatively brittle CFRPs are very sensitive to thermal cycling, and transverse cracking was found to be a dominant damage mechanism of graphite-epoxy specimens at thermal cycling at the temperature varied from −121°C to +121°C [8]. Thermal cycling slightly reduces the tensile strength and modulus of the composite, but has a more serious effect on matrix-dominated mechanical properties of the composite, such as flexure, compression and interlaminar shear [9-12]. For instance, transverse flexural stiffness and strength decreased by 25% and 34%, respectively, after 80 cycles in a simulated low earth orbit (LEO) environment with the temperature change from -70°C to 100°C [10].

In the literature, there are no reported studies with respect to the LCF behavior of Al alloys plated with multi-layered deposits. Besides, fatigue behavior of CFRPs in a purely bending mode is not studied in-depth. Therefore, an investigation of low-cycle fatigue of a multi-layered Al alloy and a CFRP is of vital importance for the design of structural components made of those lightweight materials.

2. Experimental

Solutionized and artificially aged 6061-T6 aluminum alloy consists of, wt.%: 0.4 - 0.8 Si, ≤0.7 Fe, 0.15 - 0.40 Cu, ≤0.15 Mn, 0.8 - 1.2 Mg, 0.04 - 0.35 Cr, ≤0.15 Ti, ≤0.25 Zn, other elements ≤0.05% each, 0.15% total, 95.85 – 98.56 Al. The alloy was coated with a single 12-μm or 26-μm-thick layer of electroless nickel (sets 2 and 3, respectively, Table 1). Two-layered deposits include an inner 12-μm-thick or 26-μm-thick EN and an outer 3-4-μm-thick electrodeposited Ni (EDN) layers performed in accordance with the standard SAE AMS 2424F-2010 in a nickel sulfamate bath. Three-layered coatings have an outer layer of silver (set 6) or gold (set 7) deposited on the alloy surface in accordance with ASTM B700 Type 1 Grade A Class N (thickness class 10μm) or ASTM B488 Type 1 Grade C Class 0.1μm, respectively. The samples had the gauge width and length of 10 and 32 mm, respectively.

The carbon fiber-epoxy polymer laminate used in this study was consolidated from prepreg and structural epoxy resin supplied by Hexcel Corporation (USA). The prepreg HexPly® 8552-type is an amine-cured, toughened epoxy resin system supplied with woven carbon fibers [13]. The 3-mm-thick carbon-epoxy specimens with the angle-ply configuration of ±0°/90° had the width of 16 mm, length of about 90 mm and the load span of 50 mm. The laminate has the ultimate tensile strength of 800 MPa, and elongation-to-fracture of 1.5%. As a criterion of fatigue failure of
composite (lifetime \( N_f \)), the number of cycles corresponding to a 10%-decrease in the cyclic force amplitude \( F_{\text{max}} \) was selected.

Table 1. The outer deposit kind, thickness, roughness (Ra) and Vickers microhardness (VH) as compared to those of the substrate*).

| The layer | Kind of a deposit, thickness [\( \mu m \)] and surface properties for sets 1 - 7 |
|-----------|---------------------------------------------------------------------|
| 1         | 12-EN, 26-EN, 12-EN, 26-EN, 12-EN, 26-EN, 12-EN, 26-EN              |
| 2         | 4-EDN, 3-EDN, 4-EDN, 3-EDN, 4-EDN, 3-EDN                            |
| 3         | 3-Ag, 0.2-Au                                                        |
| Ra, \( \mu m \) | 1.98, 0.75, 0.82, 0.91, 0.82, 1.22, 0.82                          |
| VH, GPa  | 1.05, 5.10, 5.10, 1.94, 1.94, 0.29                                  |

*Designations: set 6, e.g., consists of an inner 12-\( \mu m \) EN layer; an intermediate 4-\( \mu m \) EDN layer and an outer 3-\( \mu m \) silver layer.

The specimens were tested on a Model IP-2 pure bending fatigue machine with the capacity of around 50 Nm in a strain-controlled loading mode at the strain ratio \( R = e_{\text{min}}/e_{\text{max}} \) amounted to 0.1 for both materials and -1 only for CFRP, where \( e_{\text{min}} \) and \( e_{\text{max}} \) are the minimum and maximum values of the total strain, respectively. The scheme of the LCF test and a view of the sample deflection measurement are given in Ref. [14]. The sine-wave input form with the frequency of 0.4 Hz was used for Al alloy and CFRP, respectively. The plastic strain amplitudes \( \Delta e_{\text{pl}} \) varied from about 0.002 to 0.0010 for both materials. Here the \( \Delta e_{\text{pl}} \) was calculated as a difference between the maximum total strain \( \Delta e_c \) and maximum elastic strain \( \Delta e_e = TYS/E \), where \( TYS \) and \( E \) are tensile yield stress and elastic modulus, respectively. The testing environment was air at 25°C ± 2°C.

The maximum and minimum operating temperatures for the thermal cycling simulation are 180°C and −195.8°C (liquid nitrogen), respectively. Temperature increases at the rate of approximately 10°C/min and decreases approximately at 7°C/min; the duration of each of 10 thermal cycles was about 50 minutes. The reference and thermally-cycled (TC) 3 mm-thick samples had approximately the same tensile properties.

The tensile properties of uncoated and coated 6061 alloy were performed using a universal testing machine Zwick-1445 at the test speed of 2.6 \( \times \) 10\(^{-3} \) s\(^{-1} \). The hardness (average of 10 measurements) was measured using a HMV-2 Microhardness Vickers tester with a diamond pyramid under the load of 9.81 N (Al substrate), 0.98 N and 0.49 N (EN, EDN), 0.49 N and 0.25 N (Ag). The surface roughness of the alloy and coatings determined as the center-line average height \( R_a \) was measured by a Veeco Dektak 150 surface profilometer. The microscopic evaluation of the samples was performed on an optical microscope Axio Observer.A1m (Karl Zeiss, USA) and a scanning electron microscope JEOL JSM-5600 with ‘NORON’ energy-dispersive analysis system. The X-ray diffraction of EN deposit was performed by Philips X-ray diffractometer using Cu-K\(_{\alpha}\) radiation.

3. Results and discussion

3.1. Materials characterization

3.1.1 6061-T6 alloy.

The average surface roughness \( R_a \) of the substrate alloy amounted to 1.98±0.54 \( \mu m \) and was approximately two-three-fold higher than that of the deposits. Among the studied coatings, EN layer has the lowest roughness (≈ 0.8 \( \mu m \)) and the highest hardness of about 5.1 GPa (Table 1). The Vickers microhardness test (average of 10 measurements) demonstrates that hardness of an EN deposit is five times higher than that of the substrate (Table 1). Under the assumption that the EN layer consists of Ni-P binary alloy only, the average phosphorus concentration obtained by EDS analysis (10 measurements) amounted to about 9.7±1.5%P. X-ray diffraction analysis of an EN deposit showed two peaks of amorphous Ni-P phase at 20 of 44.6° and about 80.6°, which confirms the data reported by Lewis and Marshal [15] corresponding to the crystallite size ≤ 1.5 nm.

The tensile tests carried out on uncoated 6061-T6 aluminum alloy indicated an ultimate tensile strength (UTS) of 345±4 MPa, tensile yield stress (TYS\(_{0.2\%}\)) of 300±3 MPa and elongation-to-fracture (\( \delta \)) of 11.9±1.1% (Table 2). All sets of coated aluminum alloy may be divided into two groups in accordance with the deposit thickness and results of tensile tests: a group A including the alloy coated with a 12-\( \mu m \)-thick inner EN layer (sets 2, 4 and 6) and a group
B including the samples coated with a 26-μm-thick inner EN layer (sets 3, 5 and 7, Table 2). In general, the maximum strength for both groups is 4% -8% lower than in the substrate, and TYS for group A is also 6% - 9 % lower than that of the substrate. The yield properties of samples related to group B, in which the inner EN layer is two-fold greater than in group A, were practically the same as in an uncoated alloy. Elongation for the samples of the group A is slightly lower than that of an uncoated alloy (Table 2, sets 2, 4 and 6). However, the samples of the group B with a greater deposit thickness showed two-fold decrease in elongation as compared to the uncoated alloy (Table 2, sets 3, 5 and 7). In general, coated samples related to the group A are more ductile than the sample from the group B, in which the thickness of deposited layers is more than two-fold greater (Tables 1, 2).

3.1.2. Carbon fiber reinforced polymer (CFRP).

The SEM examination of the samples before and after thermo-cycling showed that thermo-cycled samples demonstrated the resin cracking originated both in the outer ply and in the bulk (Fig. 1). Besides, the fiber cracking was observed as a result of thermo-cycling, too (Fig. 1d).

3.2. Fatigue behavior of materials

A typical presentation of low-cycle fatigue test results is satisfactorily described by a well-known Coffin-Manson relation [16, 17]:

$$\Delta \varepsilon_{pl} = \varepsilon_f' N^{-c}$$

(1),

where $\Delta \varepsilon_{pl}$ is the plastic strain amplitude, $\varepsilon_f'$ is approximately equal to the true fracture strain and is called fatigue ductility coefficient, $N$ is the number of cycles to failure, ‘$c$’ is so-called fatigue ductility exponent, which, as a rule, varies for metals in a relatively narrow interval (0.4 – 0.6). The exponent $c$ reflects the ductility and hardening of metal under cycling strain. The importance of the Coffin-Manson equation consists in the possibility of its use for predicting fatigue behavior when the two parameters have been measured.

3.2.1 Fatigue properties of the 6061 Al alloy.

The $\Delta \varepsilon_{pl} – N$ diagrams for the substrate 6061 and coated alloy are presented in Fig. 2 where the curve numbers are the set numbers given in Table 1. The fatigue life of 6061 substrate dramatically decreases with an increase in the plastic strain amplitude $\Delta \varepsilon_{pl}$ from 0.003 to 0.010. It was found that the fatigue ductility exponent for an uncoated alloy amounted to 0.667, while for plated samples of sets 2 – 7, it varied from 0.404 to 0.603. Fatigue ductility

| Table 2. Tensile mechanical properties of the substrate (reference) and coated alloy |
|--------------------------------------|-------|--------|-----|
| Set No. & a group | UTS, MPa | TYS$_{0.2}$, MPa | $\delta$, % |
| 1 | 345±4 | 300±3 | 11.9±1.1 |
| 2-A | 324±6 | 272±4 | 10.1±1.6 |
| 3-B | 333±3 | 300±5 | 6.0±2.1 |
| 4-A | 331±3 | 284±6 | 10.2±1.2 |
| 5-B | 323±5 | 299±4 | 6.8±1.0 |
| 6-A | 322±4 | 281±6 | 10.0±2.0 |
| 7-B | 316±8 | 306±7 | 6.2±1.0 |

Fig. 1. Porosity between plies in the reference CFRC (a) and cracking of the resin in a thermo-cycled sample in the surface ply (b) and in the bulk (c, d) before fatigue test.
coefficient for the Al substrate amounted to 3.751 and was around an order of amplitude greater than that for coated systems represented by sets 2-7, where $\varepsilon_f$ values varied from 0.108 to 0.500 (Table 3). Thus, both fatigue and tensile tests showed a drastic deterioration of plasticity for coated 6061 alloy as compared to the substrate.

![Graph](image1)

Fig. 2. Fatigue life of Al 6061 T6 substrate (1) as compared to that for the 6061 T6 alloy coated with a single layer (2, 3); two layer (4, 5) and three-layer (6, 7) coatings depending on plastic strain amplitude.

As it can be seen from Fig. 2, deposition of a single electroless 12-μm-thick layer and, especially, a 26-μm-thick Ni layer significantly shortens the lifetime of the system 6061-T6/EN as compared to the uncoated alloy. The mean lifetime of the alloy plated with 12-μm and 26-μm-thick EN layers amounted to 13,008 and 5,535 cycles, respectively, as compared to 45,849 cycles for the substrate at the smallest plastic strain amplitude of 0.003 (Fig. 2, Table 4). Therefore, the relative lifetime $N_p/N_{p0}$ of these coated systems (sets 2 and 3) amounted only to 0.28 and 0.12 as compared to the substrate, where $N_{p0}$ and $N_p$ are the numbers of cycles to failure for the coated alloy and the substrate, respectively. An increase in the plastic strain up to 0.005 and 0.010 leads to an additional degradation of fatigue properties both for 12-μm and 26-μm-thick EN deposits.

The second 4-μm-thick EDN-layer electroplated on the electroless Ni layer (sets 4 and 5) insignificantly reduced the fatigue life of the coated alloy at a relatively high plastic strain of 0.010, probably, due to the ‘thickness effect’: the thicker the coating, the greater lifetime reduction (Fig. 2). However, at medium and low strain levels corresponding to the number of cycles more than $10^3$, a marked improvement of fatigue properties was found. For example, at $\Delta \varepsilon_{pl} = 0.003$, lifetime of sets 4 and 5 amounted to 15,014 and 9,197 cycles, respectively, as compared to 13,008 and 5,535 cycles for sets 2 and 3, which were not electroplated with EDN. Especially strong influence of EDN deposit improving the fatigue properties of a two-layered system was found for set 5 at medium and low plastic strains (Fig. 2, Table 4).

All fatigue curves obtained for a coated alloy are located

![Graph](image2)

Fig. 3. The view of an EDN layer on the lateral sample surface (set 4) after one cycle at plastic strain amplitude of 0.003 (a) and 0.010 (b).
on the left of the $N$ diagram for the substrate, which demonstrates a reduction in the cyclic longevity as a result of embrittlement of the system by EN layer (Fig. 2). It is well-known that tensile strength and hardness of an EN deposit (7% - 9%P) are almost twice higher than those of EDN and amount to 950±150 MPa and around 4.7 GPa, respectively [4]. Meanwhile, for electroplated nickel, these values amounted to 510±100 MPa and 2.0±0.3 GPa [18]. Ductility of EN is, accordingly, much less than that of EDN: elongation-to-fracture is usually equal to 1% vs. 5-30% for the coating electrodeposited in a sulfamate bath [4, 18].

In accordance to our data, the hardness value of electroplated nickel is about 1.9 GPa vs. 5.1 GPa for electroleess nickel (Table 1). It seems that at relatively minor plastic strain amplitudes, a much more ductile electroplated Ni-layer prevents crack formation on the outer surface of the coated alloy and, probably, in the interface between Ni layers. Of course, at high strains, EDN layer does not prevent the intensive cracking observed already after the first cycle contrary to that for a lower plastic strain (Fig. 3). Cracking after the 1$^\text{st}$ cycle was found at $\Delta$e$_{pl}$ = 0.010 both in the outer EDN layer and in the inner 12-$\mu$m-thick EN deposit (Figs. 3b, 4a).

The dynamics of the crack propagation in samples coated with two-Ni-layers (set 4) at the medium plastic strain of 0.005 is illustrated in the pictures corresponding to 0.2; 0.4 and 0.8 of fatigue life (Figs. 4b, 4c and 4d). It can be clearly seen that cracking occurs, mainly, at the interface between the inner electroleess nickel layer and the Al substrate. As expected, the cracks nucleated near various heterogeneities of the surface, such as a hillock-type defect (Fig. 4a, 4b).

In the uncoated alloy, the final fatigue fracture, as can be distinctly seen in Fig. 5a, has been originated at such surface defect as a pit ‘A’. The fracture process starting from a pit was dominated by the propagation of a single crack with the directions marked by arrows. In contrast to high-cycle fatigue fracture patterns, the LCF fracture surface can include the striations in the area of an initial rupture close to the sample surface [19] as demonstrated in Fig. 5b. An origin of the fracture ‘A’, a secondary crack ‘B’ and striations ‘S’ in a three-layered coated sample (set 6) are shown in Figures 6b, 6c and 6d.

In a tilted sample of set 3, very fine striations ‘S1’ in EN deposit and striations ‘S2’ in the substrate were observed (Fig. 6).
orders of magnitude at the same plastic strain of 0.0055 (Fig. 7). The similar trend was observed, e.g., for a ±45°-angle-ply carbon/epoxy laminate: under the maximum stress of 120 MPa, the lifetime of the composite increased from 10^2 to 10^5 cycles at stress ratios of -1 and 0.1, respectively [21].

The Coffin-Manson relationship (Eqn. 1) is valid both for reference and for thermally-cycled samples with fatigue ductility exponent $c$ varying in a very narrow range from -0.102 to -0.175 (Table 3). Much lower values of $c$ obtained for the laminate comparing to Al 6061 alloy ($c = -0.67$) reflect a significant higher sensitivity of the CFRP to cyclic plastic strain: a very small increase in $\Delta e_{pl}$ causes a dramatic shortening in its lifetime. An inversion point was found in the log $\Delta e_{pl}$ - log $N$ diagram which describes the fatigue behavior of reference and thermally-cycled laminates: two lines corresponding to the strain ratio of 0.1 are intersected at $N$ values of about 20,000 cycles (Fig. 7). At plastic strain amplitude exceeding 0.006 (the LCF region at $N \leq 20,000$), thermally-cycled laminate showed a slightly longer lifetime in comparison with the reference one. Thus, in the region of LCF, thermo-cycling improves fatigue resistance of the composite. The similar effect was observed earlier by Gao et al [12] for a CFRP that was thermally-cycled less than 40 times: its bending strength increased comparing to the reference one due to some increase in the cross-linking density.

### Table 4. Mean and relative lifetimes ($N_i$, $N_{i}/N_{i}^{c}$) and the standard deviation of results (SD) for uncoated (1) and coated Al alloy (2-7) as a function of plastic strain amplitude $\Delta e_{pl}$

| $\Delta e_{pl}$ | 0.003 | 0.005 | 0.010 |
|-----------------|-------|-------|-------|
| Set No. | $N_i$ | SD | $N_{i}/N_{i}^{c}$ | $N_i$ | SD | $N_{i}/N_{i}^{c}$ | $N_i$ | SD | $N_{i}/N_{i}^{c}$ |
| 1 | 45,849 | 5,223 | 1 | 28,612 | 5,969 | 1 | 6,855 | 843 | 1 |
| 2 | 13,008 | 1,120 | 0.28 | 3,234 | 473 | 0.11 | 1,266 | 23 | 0.18 |
| 3 | 5,535 | 1,476 | 0.12 | 2,299 | 467 | 0.08 | 614 | 41 | 0.09 |
| 4 | 15,014 | 1,592 | 0.33 | 6,263 | 440 | 0.22 | 902 | 37 | 0.13 |
| 5 | 9,197 | 831 | 0.20 | 3,150 | 376 | 0.11 | 407 | 21 | 0.06 |
| 6 | 9,920 | 1,546 | 0.22 | 2,957 | 546 | 0.10 | 1,137 | 71 | 0.17 |
| 7 | 6,427 | 1,338 | 0.14 | 1,757 | 541 | 0.06 | 596 | 21 | 0.09 |

3.2.1 Fatigue properties of the laminate CFRP

A very significant effect of plastic strain amplitude on the fatigue life of the composite was revealed (Fig. 7). As was obtained earlier for this material [20], a reversible loading mode significantly shortens the lifetime of the laminate in comparison with asymmetrical mode with the strain ratio of 0.1. For instance, the lifetime of 3 mm-thick laminate increases from 70 cycles for $R = -1$ to 7×10^4 cycles for $R = 0.1$ or in two

Fig. 6. The fracture surfaces with striations S1 and S2 in EN layer and in the substrate, respectively, and secondary cracks B in coating which aligned parallel (a, set 3) and perpendicular (b, set 7) to the main crack.

$\Delta e_{pl} = 0.003$; broad arrows (a) bracket 6 striations in an EN deposit.

Fig. 7. Diagram $\Delta e_{pl}$ - $N$ for the reference CFRP (1, 2) and TC- samples (3) tested at $R$ of -1 (1) and 0.1 (2, 3).
in the epoxy resin layers. Probably, in our case, at plastic strain exceeding 0.006 ($N \leq 20,000$), increasing bending strength of TC samples plays a dominant role in a higher fatigue life of the composite. On the other hand, it can be assumed that small cracks caused by thermal cycling, can help to minimize the main crack propagation due to the stress distribution between cracks and retard thereby an earlier fracture of the sample.

On the contrary, at a smaller plastic strain ($\Delta e_{pl} \leq 0.006$) and $N$ higher than 20,000 cycles, thermal cycling of the composite slightly shortens its lifetime in fatigue test (curves 2 and 3, Fig. 7). Such behavior of CFRP was found, e.g. for carbon/PEEK laminate in fatigue tests performed in a ‘push-pull’ mode, where the lifetime of TC-samples amounted only 40,000 cycles vs. 158,000 cycles for reference samples [22]. At lower plastic strains and the longer fatigue test duration, the dominant damage mechanism of the thermally-cycled composite, probably, is formation and growth of small transverse cracks resulting from thermal loading as it was reported for flat and tube carbon-fiber composites [8].

Fatigue ductility coefficient increases from 0.0171 for reference samples to 0.0367 for thermally-cycled laminate (Table 3). It points to increasing plasticity of the laminate after thermal cycling, which leads to increasing fatigue life at relatively large strains. Some damage patterns on the cross-section of a thermally-cycled sample were found (Figs. 1b, 1c, 1d). We can assume that an increasing fatigue life of thermally-cycled laminate in the region of low-cycle fatigue (less than $2 \times 10^4$ cycles) is due to the stress distribution between small cracks originated as a result of thermo-cycling, which retards the fatigue failure and prevents fast propagation of the main crack.

SEM micrographs of the failure appearing after cyclic loading performed are represented in Figure 8. In an asymmetric mode, the sample side that experienced tension exhibited, mainly, matrix and fiber cracking, while the side that experienced compression, underwent buckling, which caused delamination and cracking (Figs. 8a, 8b). In a

![Fatigue degradation patterns in a reference (a-c) and TC (d-f) samples at plastic strain amplitude $\Delta e_{pl} = 0.010$ and strain ratio $R = 0.1$.](image-url)
reversible mode, both sides of the sample underwent tension and compression and therefore, experienced matrix and fiber cracking.

Energy released at the time of the process of damage accumulation leads to the sample heating [20]. The sample temperature did not seem to change noticeably prior to failure of the first ply characterized by a drastic decrease in force, as observed for 3.3-mm thick sample in an asymmetric loading mode after about 200 cycles [20]. A weak initial reduction of the force and a small increase in the temperature due to damage accumulation can be explained by the appearance of several transverse cracks before the crack density saturation [23, 24]. Then, several damaging modes can develop simultaneously (cracking and delamination) with a final catastrophic failure of the whole sample (Fig. 8).

Summary

Low-cycle fatigue of plated 6061-T6 Al alloy and carbon-epoxy laminate was studied in a pure bending strain-controlled mode, which is more complicated than the usually reported uniaxial push-pull loading mode. It was found that the Coffin-Manson relationship is valid for the alloy as well as for the laminate with the fatigue ductility exponent ‘c’ varying in the interval 0.402 – 0.667 and 0.102 - 0.175, respectively.

The tensile and fatigue properties of the alloy coated with multi-layered deposits depend on the thickness of the electroless nickel layer (EN) that drastically decreases the ductility of the system. The fatigue life of the coated alloy is greatly shortened as compared to that of the substrate, e.g., after depositing 12-μm- and 26-μm-thick EN layers at the maximum plastic strain of 0.003, it decreased by 72% and 88%, respectively, as compared to uncoated alloy. Electroplating by the second ductile Ni layer increases the lifetime of the alloy in comparison with that coated only by the electroless Ni layer.

Thermal cycling of the composite at maximum and minimum operating temperatures of 180°C and −195.8°C, respectively, shortens the fatigue life of samples in the high-cycle fatigue range corresponding to the number of cycles N exceeding 20,000. However, in low-cycle fatigue conditions (Δε_pl ≥ 0.006 and N ≤ 20,000) the fatigue life of thermally-cycled laminates was slightly longer than that of the reference one. This behavior of thermally-cycled laminate at relatively high plastic strain amplitudes, probably, is due to the stress distribution between small cracks, which retards the fatigue failure and prevents fast propagation of the main crack.

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