Microstructure evolution of a drive shaft spline from an aero-engine fuel pump during fretting wear

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Abstract
In the process of transmitting power, the drive shaft splines of an aero-engine fuel pump often cause fretting wear due to the action of high-frequency vibration and torque load, which greatly restricts the service life and reliability of a drive shaft spline. Therefore, to understand the whole process of the fretting wear of a drive shaft spline, the microstructure of the worn surface and subsurface was characterized and analysed. The results show that adhesion, deformation, oxidation and cracking occur on the worn surface of the drive shaft spline. Plastic deformation induces dislocation multiplication that expands into subgrains and transforms into equiaxed nanocrystals. Dislocations generated by these grains in the subsequent plastic deformation will be quickly absorbed by grain boundaries, which reduces the stress concentration caused by dislocation blocks and significantly delays cracking. However, wear cracks tend to form at the interfaces of short rod-like nanocrystals near the worn surface and propagate along grain boundaries, accelerating the spalling of the material. These results provide guidance for the design and use of drive shaft splines.

1. Introduction
Involute splines play an important role in aero-engine transmission systems because of their stability, high bearing capacity and good directivity [1, 2]. To satisfy increasing requirements for the thrust-to-weight ratio, reliability and safety of aeroengines, spline performance has become extremely important. Because of its advantages of high strength, high toughness and good fatigue performance [3], 40CrNiMoA steel is now used as one of the conventional materials for drive shaft splines in aero-engine fuel pumps in China. The drive shaft in the fuel pump is connected to the driven gear by involute splines. The complex multiaxial stresses caused by high-frequency vibration and periodic torque produce fretting wear in contact areas between internal and external splines [4]. Fretting wear has two consequences: wider spline clearance that reduces the stability of transmission, and formation of cracks that accelerate spalling, increase the wear rate and lead to spline failure [5].

In research on the fretting wear of aviation splines, scholars have focused on the use of lubrication, coatings, surface quality and misalignment [6–8] to improve the surface contact conditions and resistance to fretting wear. In addition, the fretting wear depth and wear mechanisms of splines were predicted by simulations and calculations [9–11]. Unfortunately, these studies did not provide an understanding of the microstructure of worn subsurface, such as tribologically transformed structure (TTS) [12] and debris layer (DL) [13]. Due to the inconsistent behaviour of these structures in different materials, the formation of TTS and DL is still controversial. Under different fretting conditions, oxidation reactions can change the temperature of the local
contact surface and eventually lead to different substances forming DL. For example, the main component of the debris produced by fretting steel at room temperature is $\alpha$-Fe$_2$O$_3$, and in some cases, it also contains Fe$_{14}$. TTS is a hard and brittle deformed structure produced in the process of fretting wear, and they directly lead to material loss. Various mechanisms, such as dynamic recrystallization, high plastic strain and mechanical alloying, have been proposed for the formation of TTS. To date, a systematic theoretical explanation has not been developed. Therefore, to study the mechanisms of the formation of TTSs, it is helpful to first understand the evolution of the friction layer during fretting wear. During fretting, plastic deformation of the surface layer is accompanied by strong work hardening, and the hardness increases. The high hardness of TTS is the main cause of crack formation. Li studied the fretting wear of nickel-based alloys in air and pure water and indicated that delamination cracks tend to form on the interface of the TTS/wear debris layer (WDL), wear surface and WDL. Ming believed that the formation of cracks at the interface between TTS-I and TTS-II was caused by uncoordinated deformation. Tarassov pointed out that in the processes of friction and wear, the presence of nanostructures with a thickness of tens of microns near the surface had a great influence on wear. That is, the microstructure evolution caused by strong plastic deformation provided structural conditions that were favourable for the formation of wear cracks. After a crack formed, it expanded in a direction parallel to the worn surface and finally caused material to spall from the surface and become debris. During fretting, debris particles undergo repeated squeezing, crushing and oxidation and finally form nanoscale debris. In most previous studies, fretting wear tests were performed under low-frequency (0–50 Hz) experimental conditions, while the fretting wear of drive shaft splines occurs at high-frequency (200–2000 Hz) vibration. At present, the understanding of the microstructure of worn subsurface is not deep enough under the action of high frequency repeated shear. Only by fully understanding the evolution of microstructure can we fully understand the whole fretting wear process. Therefore, the purposes of this work are to understand the microstructure of drive shaft splines during fretting wear in practical engineering applications and to reveal the mechanism of fretting wear.

2. Experimental details

2.1. Sample preparation and service conditions

Figure 1(a) shows a schematic diagram of a drive shaft spline from an aero-engine fuel pump. This paper studies a damaged drive shaft spline (external spline) that was used for 900 h, as shown in figure 1(b).

Material composition: the drive shaft spline was low-alloy 40CrNiMoA steel with the chemical composition shown in table 1.
The heat treatment process of the drive shaft spline in the manufacturing process: In a vacuum heat treatment furnace, after austenitizing at 850 °C, it is quenched in oil and then held at 450–480 °C for 3 h for vacuum tempering.

Spline parameters: Drive shaft spline teeth number $z = 12$, modulus $m = 2.5$ mm, pressure angle $\alpha = 28^\circ$, effective contact length $L = 12.5$ mm, tooth surface roughness $Ra = 1.6 \mu m$, and hardness is 42.2 HRC. The hardness of the internal spline is 44.8 HRC and roughness $Ra = 0.8 \mu m$.

Service conditions: figure 1(c) shows the involute contact mode between the drive shaft spline and the internal spline. Under the action of high-frequency vibration (operating frequency in the range of 200–2000 Hz) and torque load (rated torque of 109 N.m), the contact area mainly bears forces in three directions: tangential force ($F_x$), radial force ($F_y$) and axial force ($F_z$). $F_n$ is the combined force of $F_x$ and $F_y$. The intersection of these forces is one of the main factors causing fretting wear in the contact area of the spline tooth surface. To reduce spline wear, contact areas are lubricated with grease.

Sample preparation: For observations of the worn surface morphology, a single tooth (figure 2) was obtained by cutting the damaged drive shaft spline (figure 1(b)) with wire electrical discharge machining (WEDM). Samples were extracted by a plasma focused ion beam (FIB) system (Helios G4) from the two regions with different wear degrees of the single tooth surface of the drive shaft spline. Two thin foil samples (sample T1 and sample T2) with a thickness of approximately 100 nm were prepared.

2.2. Characterization
To understand the microstructure of the worn surface and subsurface of the drive shaft spline, the overall morphology of the worn surface was first observed by laser scanning confocal microscopy (LSCM, OLS-5000). The morphologies and elemental mapping analyses of the worn surface were studied by scanning electron microscopy with energy dispersive spectroscopy (SEM/EDS, SUPRA40). Then, samples T1 and T2 were observed in detail at 300 kV using transmission electron microscopy (TEM, Tecnai G2 F30 S-TWIN). Bright field (BF), selected area electron diffraction (SAED) and high-angle annular darkfield (HAADF) were obtained in TEM mode to determine the ultramicroscopic morphology, particle size and phase structure of the samples. High-resolution TEM (HRTEM) images were acquired for phase analysis, and IFFT images were obtained by inverse fast Fourier transformation (IFFT) to confirm the dislocation distribution and density. The main chemical composition distributions were analysed by EDS mapping. In addition, sample T1 was studied by thermal field emission scanning electron microscopy (SEM, Gemini 300) and an Oxford SYMETRYS with the transmission Kikuchi diffraction (TKD) technique at 30 kV with a step size of 6 nm.

3. Results and discussion
3.1. Overall morphology of the worn surface
Figure 2 shows the overall morphology of the worn surface of the single tooth on the drive shaft spline during actual service. There is more wear at both ends of the surface of the drive shaft spline than at other positions, indicating uneven wear on the surface of the drive shaft spline. This may be due to the error in the axis alignment of the spline pair during installation [4, 26], especially in the early stage of operation. The spline load cannot be evenly distributed on the whole tooth surface due to the change in the shape of the edges at both ends of the tooth.
surface, which biases the spline load at both ends. Thus, it is no longer possible for the spline to increase the effective contact length to improve its bearing capacity. Additionally, for involute splines under the action of high-frequency vibration and torque load, each tooth undergoes a different degree of deformation along the axial direction, which leads to different deflection angles and an uneven distribution of the load along the axis. These two conditions cause a change in the contact area and contact stress at the surface of the drive shaft spline; the local contact stress alters the local subsurface microstructure. Therefore, local samples were prepared from the worn surface on the drive shaft spline for TEM analysis to further observe the changes in the subsurface microstructure.

3.2. SEM morphology of the local areas of the worn surface

Figure 3 shows SEM images and EDS results for the local areas on the worn surface of the drive shaft spline: (a) tooth top, (a1) enlarged image of the local area in (a), (b) tooth middle, (b1) enlarged image of the local area in (b), (c) tooth root, (c1) enlarged image of the local area in (c), and (d) EDS results for the corresponding areas indicated by red capital letters.

Figure 3. SEM images and EDS results for the local areas on the worn surface of the drive shaft spline: (a) tooth top, (a1) enlarged image of the local area in (a), (b) tooth middle, (b1) enlarged image of the local area in (b), (c) tooth root, (c1) enlarged image of the local area in (c), and (d) EDS results for the corresponding areas indicated by red capital letters.
microstructure changes of the two thin foil samples were compared in TEM mode, which is very important to understand the evolution of the microstructure of the wear subsurface.

During fretting wear, the fretting direction changes due to the influence of high-frequency vibration and torque load. Therefore, the direction of fretting can be determined only by the direction of the groove, i.e., at an angle of $34^\circ$–$45^\circ$ along the axial direction. The material adhesion is in the coexistence stage of elastoplastic deformation, and the deformation is large enough to destroy the microconvex bodies on the contact surface, that is, material transfer occurs. The EDS results confirm this. Material transfer is an important sign of adhesive wear. The formation and propagation of surface cracks accelerate the spalling of the material, and large and small pits form after spalling. Under high-frequency repeated shear, this process occurs repeatedly, resulting in delamination. The spalled material is squeezed, crushed and oxidized under the action of mechanical force and eventually forms submicron or nanoscale oxide particles [23–25]. These oxide particles are the main source of abrasive wear, and grooves are the most intuitive manifestation of abrasive wear. Comprehensive analysis shows that the main damage mechanism of the worn surface of the drive shaft spline is adhesion, deformation, oxidation and cracking.

### 3.3. Characterization of the microstructure of the worn subsurface

Figure 4 shows the microstructure and EDS results of sample T1 in TEM mode. The complex layered structure in the overall HAADF image (figure 4(a)) consists of a protective layer (Pt), a debris layer (DL), a tribologically transformed structure (TTS) and a general deformation layer (GDL), consistent with the results of previous studies [13, 27]. The Pt layer was deposited by FIB technology to protect the worn surface during sample preparation. The thickness of the DL is approximately 60 nm, and the thickness of the TTS is approximately 950 nm. Figure 4(b) is the HAADF image of the area enclosed by the red dotted rectangle in figure 4(a); it shows the structure of the DL and TTS at a higher magnification, while figures 4(b1)–(b3) shows the elemental distributions of Fe, O and C in this area. The DL is mainly composed of Fe-rich oxides, which is consistent with the EDS results for the oxide nanoparticles in figure 3. The oxygen content in the TTS is extremely low, and there is no obvious oxide formation. In addition, no significant cementite was observed in the TTS.

Figure 5 shows the local microstructure of the GDL and TTS (as labelled in figure 4). Figure 5(a) is a local BFTEM image of the GDL. The grain interface in this area is clearly visible and has a distinct lamellar structure. Figure 5(b) is the HRTEM image of the local area (position A) in figure 5(a). The IFFT image is obtained by inverse fast Fourier transformation of the selected area, and the dislocation density inside the grain in the GDL is calculated to be $1.56 \times 10^{13} \text{ cm}^{-2}$. The calculation method uses the total number of dislocations in the area divided by the total area of the area. The SAED image of the GDL in figure 5(c) shows that the GDL has a ferrite ($\alpha$-Fe) structure. Figure 5(d) is a local BFTEM image of the TTS, where the grain interface is not clear and the lamellar structure disappears. Figure 5(e) shows the HRTEM image of the local area (position B) in figure 5(d) and the IFFT image of the corresponding selected area. The dislocation density inside the grain in the TTS is calculated to be $1.25 \times 10^{12} \text{ cm}^{-2}$. Compared with the GDL, the dislocation density in the TTS is reduced, indicating that in the process of shear deformation, when the dislocation density accumulates to a certain extent, it expands to form subgrain boundaries, resulting in the reduction or annihilation of dislocations. Figure 5(f) shows the SAED image of the TTS, confirming that the TTS has a normal ferrite ($\alpha$-Fe) structure.

![Figure 4](image_url) (a) Overall HAADF image of sample T1 (as labelled in figure 3), (b) HAADF image of the area enclosed by the red dotted rectangle in (a) with higher magnification, and (b1)–(b3) distributions of the main elements (Fe, O and C) corresponding to (b).

![Figure 5](image_url) 5
Figure 6 (a) illustrates the local microstructure of the DL/TTS interface (as labelled in figure 4). The DL is composed of particles ranging in size from 5 to 20 nanometres. Figure 6 (b) and figure 6 (c) are the HRTEM and SAED images at the DL/TTS interface, respectively; they confirm that the oxide nanoparticles in the DL are composed of Fe₂O₃ particles.

Figure 7 shows the TKD result of the top area of sample T1 where figure 7 (a) shows the microstructures of the selected area, and the grains of the TTS and GDL are clear. The martensite grains near the worn surface are refined very obviously, and their orientation is almost completely parallel to the worn surface. The TTS is composed of equiaxed grains with a small aspect ratio, ranging from 30 to 120 nm, while the GDL is composed of short rod-like grains, ranging from 170 to 400 nm. The change in grain size from the GDL to the TTS is similar to the gradient nanostructures (GNS) obtained by surface mechanical rolling treatment (SMRT) [28]. Some grain boundaries in the grain orientation diagram (figure 7 (b)) may not be completely resolved because the grains are too small, and therefore, black spots appear at these grain boundaries. Compared with the GDL, the grain sizes in the TTS are much smaller and their orientations are random, while the grains in the GDL are larger, have short rod-like shapes and have 〈101〉 orientations. The change in grain orientation indicates that grains in the TTS rotate during the deformation process, i.e., that the deformation mechanism near the surface layer is mainly based on coordinated sliding deformation between the equiaxed nanocrystals, while the deformation structure in the GDL is still caused by the rheology produced by shear deformation. In addition, comparing figures 7 (a)
and (b) shows that subgrain boundaries are present in some grains in the TTS. According to the diagram of residual strain (figure 7(c)), the residual strain at the subgrain boundaries in the TTS is relatively high, which is a manifestation of dislocation multiplication and expansion to form subgrain boundaries. The phase distribution in this area (figure 7(d)) is ferrite and tiny amounts of cementite (Fe$_3$C). Because the Fe$_3$C domains are too small, they cannot be obtained by SAED characterization.

Figure 8 shows the microscopic morphology and EDS results of sample T2 in TEM mode. According to the HAADF image (figure 8(a)), the multilayer structure of the worn subsurface is composed of a Pt layer, a DL and a GDL. The thickness of the DL is approximately 1720 nm, and the GDL is composed of short rod-like nanocrystals with a size range of 30–100 nm, which is inconsistent with the results for sample T1. In addition, cracks were observed in the DL and near the DL/GDL interface. SAED shows that DL has a haematite ($\alpha$-Fe$_2$O$_3$) structure. Figure 8(a1) shows the distributions of the main elements (Pt, O, Fe, and C) in (a), (b) HAADF image of the area enclosed by the red dotted rectangle in (a) at higher magnification, (b1) distributions of the main element (O and Fe) in (b); and (b2) EDS line-scan along the line in (b).

Figure 9 shows the BFTEM image at the DL/GDL interface (as labelled in figure 8(b)), while figures 9(b) –(e) corresponds to the HRTEM images of the positions marked by the red capital letters in figure 9(a).
IFFT image is obtained by inverse fast Fourier transformation near the crack area. According to the dislocation distribution, the dislocation density around the crack area (figures 9(c)−(e)) is higher than that at the DL/GDL interface (figure 9(b)), indicating that some of the dislocations generated during the deformation process migrate to the worn surface, which greatly reduces or annihilates the dislocations. The rest of the dislocations easily migrate to the interfaces of short rod-like nanocrystals near the worn surface and form dislocation tangles (DTs) due to the hindrance of the grain boundaries. With the further increase in accumulated strain, these dislocations accumulate instead of quickly separating from the tangle, resulting in stress concentration and the formation of wear cracks. That is, even without any inclusions, wear cracks easily form at the interfaces of short rod-like nanocrystals near the worn surface due to the large stress concentration caused by dislocation blocks and propagate along grain boundaries.

3.4. The evolution of the microstructure of the worn subsurface

Figure 10 shows the model of the evolution of the microstructure of a worn subsurface during fretting wear under high-frequency vibration conditions. The model illustrates the whole process from the plastic deformation at the beginning of fretting wear to the formation of a debris layer. The details are as follows:

In the process of fretting wear, the stress and strain on the contact surface are maximum and gradually decrease with increasing depth [29], creating mechanical conditions for the evolution of the martensite structure. At the initial stage of fretting, because the internal and external spline tooth surfaces are not absolutely smooth, the contact of the microconvex bodies in the local area causes the oxide film on the surface to be broken, and the surface layer undergoes microplastic deformation. As fretting continues, the actual contact area increases, and plastic deformation induces multiplication of grain internal dislocations, constant accumulation of strain and surface activation of many dislocations. These dislocations move and interact, easily forming dislocation tangles (DTs) and high-density dislocation walls (DDWs) in the deformed microstructure. This will not only refine the lamellar martensite, but also fracture or even dissolve the cementite [30, 31].

In the range of 0–950 nm from the outermost surface, the martensite refinement is very obvious. When the near-surface strain accumulates to a certain extent, the high-density dislocation structure region formed by the DTs and DDWs transforms into subcells and small-angle subgrain boundaries. At this time, the lamellar martensite divides into subgrains of different sizes, and the dislocations within the grains are greatly reduced or annihilated. The small-angle grain boundaries gradually transform into large-angle grain boundaries, and new grains form on the submicron or nanoscale. When the refinement process continues to a certain degree, equiaxed nanocrystals with a small aspect ratio form. These grains constitute the tribologically transformed
structure (TTS); its mechanism of formation is similar to that of nanostructures prepared under surface mechanical attrition (SMA) treatment conditions [32].

During fretting wear, cracks easily form in the DL when debris accumulates on the local contact surface. This is mainly attributed to the fact that DL is composed of brittle phases ($\alpha$-Fe$_2$O$_3$) that are prone to cracking under mechanical forces [33]. However, wear cracks tend to form at the interfaces of short rod-like nanocrystals near the surface due to the large stress concentration generated by dislocation blocks [34–36] and propagate along grain boundaries. The direction of propagation is approximately parallel to the contact surface, i.e., the speed of wear is greater than the speed at which the microcracks begin to propagate to the substrate, so cracks that are perpendicular to the surface and propagate to the matrix cannot be formed [37]. Therefore, the damage to the surface of the drive shaft spline is mainly controlled by wear. In addition, the intrusion of oxygen increases the crack propagation rate and accelerates material loss.

As above, under the action of high-frequency repeated shearing, the dislocations generated by the subsequent plastic deformation of the equiaxed nanocrystals near the surface layer are quickly absorbed by the grain boundaries. At this time, the deformation is mainly based on coordinated sliding deformation between equiaxed nanocrystals, which reduces the microstress concentrations caused by the dislocation blocks and considerably delays the formation of cracks. That is, the equiaxed nanocrystals formed on the worn subsurface can improve the fretting wear resistance. If the worn subsurface is composed of short rod-shaped nanocrystals, wear cracks will easily form at the interface of short rod-shaped nanocrystals under the action of uncoordinated deformation. Subsequent interconnections with cracks generated on the worn surface may lead to TTS spalling, thereby accelerating fretting wear and reducing service life. The debris formed after TTS spalling plays a key role in fretting wear performance.

Debris mainly comes from two sources. First, debris forms at the initial stage of fretting when the internal and external spline contact destroys the surface film, and the local area of the surface is scratched. Under the action of adhesion and fatigue, part of the contact surface strips or tears the substrate during the fretting wear process to form submicron or nanoscale debris. Second, the formation and propagation of wear cracks accelerate the spalling of materials and promote the formation of debris. With the high-frequency microscale amplitude reciprocating movement, these debris particles are continuously squeezed and crushed during fretting. The contact surface temperature rises, and a large amount of oxygen invades the contact surface of the drive shaft spline and reacts with the debris, forming oxidized debris particles. With the increasing number of fretting cycles, these oxidized debris particles are re-extruded and reoxidized to form nanoscale Fe$_2$O$_3$ particles. However, the presence of debris particles affects the subsequent fretting wear [38]. Some of the debris particles fill pits and are pressed onto the contact surface, forming a dense debris layer (DL) and reducing the wear rate. The rest of the debris particles are expelled dynamically from the contact area and cause abrasive wear on the surface of the drive shaft spline, which increases the wear rate and ultimately leads to material loss. Therefore, it is also very important to study the evolution mechanism of debris for fretting wear performance. In fact, the
distribution of nanoscale Fe₂O₃ particles on the contact surface of the drive shaft spline is not uniform and the number is small, so it cannot contribute to self-lubrication [35] and thus does not slow down fretting wear.

4. Conclusions

LSCM, SEM, FIB, TEM and TKD were used to characterize the microstructure of the worn surface and subsurface of a drive shaft spline from an aero-engine fuel pump. The mechanisms for the formation of the TTS and DL were clarified, and a model of the evolution of the microstructure of the fretting wear subsurface under high-frequency vibration was established. According to the detailed analysis of the experiment, the following main conclusions were drawn:

(1) The worn surface of the drive shaft spline underwent uneven wear, and adhesion, deformation, oxidation and cracking occurred. There was a complex layered structure on the worn subsurface, namely, the DL, TTS and GDL. Among them, DL is composed of Fe₂O₃ particles at the nanoscale (5–20 nm), and TTS is composed of equiaxed nanocrystals (30–120 nm). Due to the different microstructures in different regions of the worn subsurface, the GDL may be composed of short rod-like grains at the submicron scale (170–400 nm) or nanoscale (30–100 nm).

(2) Under the action of high-frequency repeated shearing, plastic deformation of the worn subsurface and many dislocation multiplications were induced. These multiplied dislocations expand into subgrains and gradually transform into large-angle equiaxed nanocrystals. The dislocations generated by these grains in the subsequent plastic deformation were quickly absorbed by the grain boundaries. At this time, the deformation was mainly based on coordinated sliding deformation between equiaxed nanocrystals, which reduced the microstress concentration caused by dislocation blocks and considerably delayed the formation of cracks. Therefore, equiaxed nanocrystals produced by the worn subsurface are beneficial for resisting fretting wear.

(3) Wear cracks tend to form at the interfaces of short rod-like nanocrystals near the worn surface due to large stress concentrations caused by dislocation blocks, and they propagate along grain boundaries. This shows that the short rod-like nanocrystals formed on the worn subsurface are not good for resistance to fretting wear.

(4) Due to the small quantity and uneven distribution of oxidized nanoparticles produced on the wear surface, they cannot play a self-lubricating role and thus will not slow down fretting wear.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

Conflict of interest

The authors declare that they have no conflict of interest.

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