Influence of cutting speed on fretting wear properties of UVAM-processed NAB alloy

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1. Introduction

Surface engineering has played an essential role in improving the friction and wear properties of materials, with significant progress in tribological optimisation and improvement [1], [2], [3]. Friction and wear processes are complex and depend on the synthesis of many material properties during contact [4]. With the increasing demand for high wear resistance of materials, the improved tribological behaviours of functional surfaces have attracted extensive research attention. The surface properties of materials have a direct impact on the friction and wear properties [5]. Surface engineering focuses on strengthening the surface properties of bulk materials, effectively improving their tribological properties. In addition, surface engineering enhances the material through functional surface layers to increase wear resistance, hardness, and strength [6, 7]. Surface engineering techniques can augment materials with excellent anti-wear and anti-friction properties.

Many surface engineering methods have been used to improve tribological properties including excimer laser machining [8], electrical discharge machining [9], thermal spraying [10], electroplating, coating [11], vapour deposition [12], and optimisation of the machining process [13]. In recent years, new coating materials, high-energy beam surfacing technologies, and nano-surface engineering have developed rapidly. For optimization of the machining process using nano surface engineering technology, UVAM technique offers the advantages of extended tool life, improved surface finish, and high accuracy [14, 15]. It is known that different microstructure grain sizes, precipitate distributions, morphologies, and contents have different influences on the tribological properties. In the subsurface layer of a UVAM-processed surface, the microstructures formed during the UVAM process have a significant impact on surface properties such as hardness and strength.

The effects of the UVAM process have been previously studied. Amini et al. [16] reported the superior tribological properties and hardness of Al7075-T6 alloys processed using ultrasonic vibration-assisted turning compared with those processed using conventional turning. Shen et al. [17, 18] reported promotional effects on the sliding tribological properties of aluminium alloy surfaces. Zheng et al. [19] investigated the tribological performance of TC4 alloy surfaces machined using UVAM. There are few investigations that refer to the impacts of microstructures in the nanoscale layer with UVAM processing. The microstructures of the subsurface layer, including the grain size, precipitate distribution, morphology, and content influence the tribological properties. Moreover, the UVAM cutting speed has significant effects on the microstructure in subsurface layer.

NAB alloys have been extensively used in navy equipment, including valves, controllable-pitch propellers, and pumps [20]. A cast NAB alloy is composed of complex phases in the binary system of copper and aluminium, resulting in diverse properties and applications. The copper-rich Widmanstätten α phase, intermetallic phases (κ, κi, κii, and κiii), and the retained β phase (β' phase) constitute the main phases in cast NAB alloy [21, 22, 23]. The aluminium produces a low stacking fault energy, which indicates a greater propensity to form nanotwins in the deformation process [24].

The introduction of defects into the material can hinder dislocation movement, promoting the strengthening of mechanical properties [25]. Traditional strengthening mechanisms inevitably affect the plastic deformation capacity, usually leading to a reduction in plasticity and toughness [26]. A nano-twinning structure can significantly improve the strength of alloys without compromising their plasticity. In traditional strengthening methods, it is difficult for dislocations to slip along grain boundaries owing to their inherent disorder. Accordingly, grain boundaries have a limited capacity to accommodate dislocations; plasticity decreases with grain boundary strengthening [27, 28]. Twin boundaries can block dislocation movement and absorb dislocations to accommodate large plastic deformations. The unique mechanical properties of nanocrystalline twin materials come from the interaction of dislocations and twinning, which is typically the main strengthening mechanism when the twin lamella thickness is at the nanoscale [29, 30].

The influence of such microstructures on the tribological properties of a subsurface layer processed by UVAM remains unknown. This study reports our research investigating the effects of cutting speed on the microstructures and fretting wear properties of NAB alloys, and the strengthening mechanisms at different cutting speeds.
2. Experimental procedures

2.1. Materials and UVAM process

In this study, the experimental material is a copper-based alloy with a composition of 10Cu-4.5Ni-9Al-4.2Fe-3.5V-1.2Mn (wt%). UVAM device was assembled with vibration components to generate vibrations during operation, and it is superior to conventional milling equipment. The vibrations lead to trajectory changes in the milling tool in processing the sample surface. An ultrasonic device (Fig. 1) was fixed at the vertical machining centre of KVC1050N. Alternating current signals were converted to ultrasound electrical oscillation signals using an ultrasonic generator. With inverse piezoelectric effects, the piezoelectric ceramic transducer converted electrical oscillation signals into mechanical vibrations that were amplified through movements of the horn and cutter. Vibrations were generated and transferred to the milling tool according to the current input.

The feed direction of the tool is perpendicular to the vibration direction. The milling tool is an 8-mm diameter solid carbide end mill with four edges. To exclude the effect of tool wear, all samples were processed using the same tool. The cutting depth was 8 mm for all manufacturing processes. To investigate the effects of cutting speed on the torsional fretting wear properties of NAB alloys, three cutting speeds were used (90 m/min, 120 m/min, and 150 m/min), with a feed rate of 80 mm/min and an ultrasonic vibration current of 200 mA. Three samples were machined for each UVAM process (Table 1).

Table 1. UVAM processing schemes.

<table>
<thead>
<tr>
<th>Surface reference</th>
<th>Processing parameters</th>
<th>Spindle speed (rpm)</th>
<th>Cutting speed (m/min)</th>
<th>Feed rate (mm/min)</th>
<th>Ultrasonic vibration current (mA)</th>
</tr>
</thead>
<tbody>
<tr>
<td>UVAM3600</td>
<td></td>
<td>3600</td>
<td>90</td>
<td>80</td>
<td>200</td>
</tr>
<tr>
<td>UVAM4800</td>
<td></td>
<td>4800</td>
<td>120</td>
<td>80</td>
<td>200</td>
</tr>
<tr>
<td>UVAM6000</td>
<td></td>
<td>6000</td>
<td>150</td>
<td>80</td>
<td>200</td>
</tr>
</tbody>
</table>

2.2. Fretting tests

Each sample was machined for three parallel parts to be adaptive for the fretting wear test rig [31], [32], [33], as demonstrated in our previous study. A torsional wear test rig was used to conduct the fretting wear tests (Fig. 2). A two-phase hybrid stepping motor which resolution is 0.02° controlled the angular displacement amplitude (θ) and the rotation speed of the lower holder. A torque sensor was used to record the friction torque of the entire wear process in real time. In naval engineering facilities systems, NAB alloy is often used in the lower pair in friction pairs of blade bearings. Severe fretting is inevitable in blade bearings, as it is a key part of a controllable pitch propeller [31], [32], [33]. To simulate the actual working conditions as much as possible, with oil lubrication, all fretting tests were conducted for 40,000 cycles; the frequency of the stepping motor was 2 Hz. The test conditions for all samples were θ = 1.5° and F = 86 N. The lubricated oil is International Organization for Standardization (ISO) VG46 and its kinematic viscosity at 40 °C is 46 ± 4.6 mm²/s.
The mechanical properties of friction pairs are listed in Table 2. In the test rig, the NAB alloy was used as the lower specimen; the upper specimen was forged 42CrMo4 steel. The hardness of the 42CrMo4 steel was 220 HB, and the roughness was 0.2 μm. Average values of the results from the three tests were used for subsequent analyses.

<table>
<thead>
<tr>
<th>Material</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elasticity modulus (MPa)</th>
<th>Hardness (HB)</th>
<th>Poisson’s ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>NAB alloy</td>
<td>250</td>
<td>650</td>
<td>121,000</td>
<td>127</td>
<td>0.33</td>
</tr>
<tr>
<td>42CrMo4</td>
<td>550</td>
<td>800</td>
<td>212,000</td>
<td>220</td>
<td>0.3</td>
</tr>
</tbody>
</table>

2.3. Microstructure analysis

An XRD-7000S X-ray diffractometer (Cu Kα) was used to detect the phase constituents. A JSM-7600F field-emission scanning electron microscope equipped with electron backscattered diffraction (EBSD) and energy dispersive X-ray spectroscopy (EDX) were used to observe the microstructures. There-dimension parameters of surface topography were measured by ultra-depth three-dimensional microscope (OLYMPUS, DSX510). A focused ion beam (FIB) with a Quanta 3D FEG SEM was used to prepare the transmission electron microscope (TEM) specimen; the tested surface of the TEM specimen was parallel to the machined surface. TEM experiments were performed using a Tencai G2 F30 instrument.

2.4. Microhardness test

The cross-sections of all samples were tested using a microhardness tester (MH-5) before the fretting tests were performed. The tested surfaces were polished to a non-scratched surface using emery paper. From the edge, the interval between two adjacent indentations was 50 μm, and the total distance from the edge was 500 μm. Each hardness value was averaged from three tests.

3. Results and discussion

3.1. Fretting results and discussion

3.1.1. T–θ curve

The T–θ curve is critical for demonstrating the running process during fretting [31]. Overall, T–θ curves can be broadly divided into three categories: linear, elliptical, and parallelogram, which correspond to fretting running states of partial slip, mixed slip, and gross slip [31].

The normal loads Fn and θ are the two main factors in the variation of the T–θ curve [32]. To simulate actual naval working conditions, the test conditions for all samples were θ = 1.5° and Fn = 86 N.

The T-θ curves are shown in Fig. 3. They exhibit a parallelogram shape, demonstrating that the systems ran in the gross slip regime. Friction pairs are composed of metals, with a large contact area that exhibits high contact stiffness. In Fig. 3, the inclined part of the curve in a single fretting cycle is influenced by the system stiffness.
The interface in the inclined part presents a partial slip regime, and the friction is static friction. As the angular displacement increases, the static friction begins to increase until the maximum static friction is reached. Relative motion occurred between friction pairs and the dynamic friction remained constant.

![Image: T-θ curve for all samples as a function of number of cycles.](Fig. 3)

3.1.2. Friction torques

The friction torque can be explained based on the average amplitude $T_i$ of a $T-\theta$ curve (Fig. 4) [33,34].

![Image: Average amplitude $T_i$ in T-θ curve in each fretting cycle.](Fig. 4)

The friction torque curve consists of four stages (running-in stage, rising stage, decreasing stage and stable stage) during fretting process, as shown in Fig. 5. The friction torque curves of all samples show the same variation tendency. At the running-in stage, with the protection supplied by the surface oxide film and impurities, the friction torque was low. With the destruction of the oxide film and work hardening caused by reciprocating motions between counter surfaces, the friction torque curve increased significantly, accompanied by small fluctuations in the rising stage. As with the progress of the fretting tests, in the decreasing stage, a third-body wear debris layer was formed on the wear scar during fretting from the wear debris and the oxide of wear debris. Between the contact interfaces of the friction pair, the third-body wear debris layer was able to protect the surface of the friction pair to some extent and reduce friction. The friction torque began to decrease after reaching the peak.
In the stable stage, the wear debris generated during fretting wear reached a dynamic balance with the removal of wear debris. The contact surface morphology and composition changed constantly. Under the action of three-body motion, the friction force reached a relatively stable state of continuous fluctuation and the fluctuation range was not large in the stable stage.

Fretting tests were performed under oil lubrication for 40,000 cycles. The last 10,000 cycles in the tests were considered to be the stable stage. From cycle 30,000 to cycle 40,000:

\[
T_{\text{sta}} = \frac{\sum_{i=1}^{10000} T_i}{10000}
\]  

(1)

\(T_{\text{sta}}\) refers to the mean friction torque in the stable stage (Fig. 6). It is observed that \(T_{\text{sta}}\) decreases as the cutting speed increases. The \(T_{\text{sta}}\) of UVAM6000 is 29.5% less than that of UVAM3600.

### 3.1.3. Accumulated dissipated energy

\(E_T\) is the accumulated dissipated energy during fretting, and represents the accumulation of integrals in each fretting cycle. The dissipated energy has an impact on the intensity of wear; the \(E_T\) measured by the testing machine helps to identify wear characteristics. \(E_T\) can be defined as [19].

\[
E_T = \sum_{i=1}^{n} E_i = \sum_{i=1}^{n} \int_{\theta} T^* \theta d\theta
\]  

(2)
The accumulated dissipated energy is strongly related to $\theta$ and $T$. In this study, with the same fretting conditions, the friction torque has a considerable impact on the ET. The variation trend of ET with UVAM parameters is consistent with that of the friction torque, as shown in Fig. 7.

![Fig. 7. Accumulated dissipated energy of each sample.](image-url)

The ET in UVAM3600 was 26.6% greater than in UVAM6000. The variation trend of ET for each sample is the same for the cutting speed as for the friction torque, as ET is the work of friction torque.

### 3.1.4. Worn surface observations

Normal loads, angular displacement amplitudes, and lubrication conditions are the three main factors contributing to the torsional fretting wear failure [30]. As depicted in Fig. 8, typical morphologies of worn surfaces were indicated. The dominant wear mechanism was abrasive wear. Ridges and scratches occurred along the directions of relative movement of the counter surfaces.

![Fig. 8. Typical morphologies of wear scars.](image-url)

Oxidation wear, cracking, and delamination are the three main wear mechanisms [30]. Lubrication decreased the shear stress to a certain extent when fretting, cracks, and delamination were not observed. The emergence and propagation of fatigue cracks were restricted. The wear debris participated in the load-bearing process, leading to a decrease in the apparent contact area. The wear debris layer was between the upper specimen and the lower specimen, reducing the direct contact between the upper specimen and the lower specimen. The experimental results indicate that plastic deformation and abrasive wear were the wear mechanisms in the oil-lubricated condition. Slight oxidation wear was also observed. EDS was used to characterize the elemental composition of the oxides in sample UVAM6000, as shown in Fig. 9. The oxygen content was slightly higher after fretting.
XPS was conducted for a better chemical analysis of the wear scars, as shown in Fig. 10. Standardised data were used to calibrate the spectra for each product [31]. The XPS mode and the Shirley method were used to create baselines. The curves were smoothed using the Savitzky–Golay method, and the peaks were fitted in the XPS spectra using the statistical method; the Al 2p core level spectra are shown in Fig. 10(a). The peaks at 74.7 eV and 75.7 eV match the centre of the Al 2p peak in Al2O3 and the Al 2p peak in Al2O3/Al, respectively, indicating that oxides were the active constituents in the worn surface. In Fig. 10(b), peaks at 932.6 eV, 943.0 eV, and 952.3 eV correspond to Cu 2p3/2 in Cu, Cu 2p3/2 in CuO, and Cu 2p1/2 in CuS, respectively. In addition to Al2O3, Fig. 10(c) shows that CuO and Al2O3 were formed, as evidenced by the centre of the O 1S peak in CuO at 530.0 eV and the centre of the O 1S peak in Al2O3 at 532.3 eV; this is consistent with the observations in Fig. 10. In Fig. 10(d), the carbon observed on the worn surface may stem from carbide impurities in the lubricant.
The worn surface area was analysed to further investigate the wear mechanisms of the UVAM samples. Greater friction torque contributed to more severe abrasive wear during the fretting process. As shown in Fig. 11, the worn surface area in UVAM3600 was 5.6% higher than that in UVAM6000; the amount of worn surface area decreased with an increase in the cutting speed, with the same trend as the friction torque.

The experiment results of the friction torque curve, the dissipate energy $E_T$, $T-\theta$ curve, worn surface observations and the mean friction torque in the stable stage $T_{sta}$ indicate the same tendency with the worn area, demonstrating the positive effects of the cutting speed on fretting wear properties.
3.2. Hardness results
Fig. 12 shows the microhardness distribution of the cross-sections at different cutting speeds. UVAM-processed NAB alloys showed increased microhardness with increasing cutting speed, contributing to the high fretting properties of UVAM-processed NAB alloys. The highest microhardness (198 Hv) was obtained in the UVAM6000 sample, and it was approximately 11% greater than that of the UVAM3600 sample (178.6 Hv). Based on the Archard theory [35],

\[ w = K s \frac{P}{H} \]  

where \( w \) represents the wear volume; \( P/H \) is the ratio between the normal load \( P \) and the hardness \( H \); \( K \) is the wear coefficient for friction pairs; \( s \) is the sliding distance. Thus, the hardness has a positive effect on the tribological performance.

![Graph showing hardness distribution](image)

Fig. 12. Hardness distribution in cross-section.

3.3. Microstructure characterizations and discussion
3.3.1. SEM observations
The typical microstructure is shown in Fig. 13. UVAM-processed samples usually consist of an \( \alpha \) phase, \( \kappa \) phases (\( \kappa I, \kappa II, \kappa III, \) and \( \kappa IV \)), and retained \( \beta \) phases (\( \beta' \) phase). With the shape of a flower, phase \( \kappa I \) is rich in Fe. Similar to phase \( \kappa I \), with a high content of Fe, \( \kappa II \) is smaller in size. Phase \( \kappa III \) has a lamellar morphology with a high Ni content. Phase \( \kappa IV \) exhibits a small spherical structure with a high Ni content. The retained \( \beta \) phase (\( \beta' \) phase), with a structure similar to that of martensite, is considered to have positive effects on the hardness of NAB alloys.
SEM images of the microstructures on the edge of a cross-section near the surface are shown in Fig. 14. The surface engineering methods of UVAM have a strong impact on the distribution and morphology phases in NAB alloys. In the UVAM3600 sample, the α phase, κ phases, and β phases were unevenly distributed. The amount of κ phase was relatively small; the κ phase had a size of approximately 0.4–0.6 μm. In the UVAM4800 and UVAM6000 samples, the κ phases are finer, at approximately 300 nm and 100 nm, respectively. More β phases were observed in the UVAM4800 and UVAM6000 samples. Elongated κ phases along the feeding direction were observed, and almost all the κ phases were stretched in the same feeding direction, indicating that the plastic deformation caused by the UVAM process alters the morphologies of the phases. Fine κ phases with a homogenous distribution were observed in the UVAM6000 sample.

3.3.2. EBSD characterizations

Fig. 15 shows the EBSD results of the UVAM-processed specimens at cutting speeds of 90 m/min, 120 m/min, and 150 m/min. At a step size of 0.5 μm, α grains and κ phases (κI, κII, κIII) are detected; the martensitic β phases cannot be completely detected. The black spot representing the undetected area indicates the fragmentation of α grains with severe deformation as the cutting speed increases. According to the EBSD characterisations, the average grain sizes for UVAM3600, UVAM4800, and UVAM6000 were 8.2 μm, 7.4 μm, and 6.0 μm, respectively.
3.3.3. XRD patterns

The XRD patterns of the UVAM-processed samples at cutting speeds of 90 m/min, 120 m/min, and 150 m/min are shown in Fig. 16.

The Scherer Eq. [36] illustrates the relationship between the grain size D and full width at half maxima βD (peak broadening):

$$\beta D = \frac{K\lambda}{D \cos \theta}$$  \hspace{1cm} (4)
where \( \lambda \) and \( K \) are the wavelength of the X-ray source and the shape factor, respectively. Compared to the UVAM3600 sample, the diffraction peak of \( \alpha(\text{Cu}) \) broadened as the cutting speed increased, demonstrating the size refinement of \( \alpha(\text{Cu}) \) during the UVAM process. Larger (Fe, Ni)Al peaks were observed as the cutting speed increased, implying a higher content of \( \kappa \) phases.

### 3.3.4. TEM observations

A high density of dislocations and stacking faults were observed in the UVAM3600 and UVAM4800 samples. Fig. 17 shows the TEM results for the UVAM3600 and UVAM4800 samples; Fig. 18 shows the TEM results for the UVAM6000 sample. A high density of dislocations and fine uniformly distributed \( \kappa \) phases are observed in Fig. 17, Fig. 18. Most of the fine \( \kappa \text{III} \) phases are surrounded by high densities of dislocations, and partial dislocations are observed in the interior of the fine \( \kappa \) phase. A high density of stacking faults was also observed.

![TEM images](image)

**Fig. 17.** (a) Typical morphology of TEM specimen; (b) HRTEM image of typical morphology in UVAM3600 sample; (c) Typical morphology of TEM specimen; (d) HRTEM image of typical morphology in UVAM4800 sample.
3.4. Analysis of strengthening mechanisms

3.4.1. Role of κ phases

According to the EBSD results (Fig. 15), as the cutting speed increases, more fine particles are intensively distributed in the matrix. In the SEM observations (Fig. 14), a high content of rich κIV phases is observed in the UVAM4800 and UVAM6000 samples. It can be inferred that the well-distributed κIV phases are finer and more intensive than those in the UVAM3600 sample, which is consistent with the SEM observations and XRD patterns. During UVAM processing, the surface temperature increased as the cutting speed increased from 90 m/min to 150 m/min, leading to the fragmentation of coarse κ phases into fine κIV phases. Under the mechanical actions of processing, the κ phases were also sheared into finer κIV phases. Through pinning, the well-distributed fine κIV phases prevented the growth of grains, and played a significant role in second-phase strengthening.

3.4.2. Effects of dislocations and stacking faults

Fig. 17 shows the TEM images and high-resolution TEM (HRTEM) images of UVAM3600 and UVAM4800. Fig. 18 shows the TEM results for UVAM6000. A high density of dislocations is observed in all samples, which has strong effects on the work hardening of the NAB alloys, contributing toward the improvement in hardness. In the HRTEM images, dislocation entanglements are also observed in UVAM3600 and UVAM4800. It is favourable for the dislocation to split into parts with wide stacking fault bands to produce stacking faults with a high density. These stacking faults acted as a barrier for all dislocations to cross-slip or climb, improving the strength of alloys. Fine grains with radii of 50–200 nm were observed in UVAM6000. These fine phases stretched in almost the same direction and were surrounded by high densities of dislocations. The fine phases may have come from the
recrystallisation; regions with high densities of dislocations contain high storage energy, providing sufficient energy for the nucleation and growth of grains.

3.4.3. Martensite nanotwins

In Fig. 18(c) and (d), the electron diffraction in the selected area shows martensite nanotwins with a thickness of approximately 30 nm. During the plastic deformation process in metals and alloys, there are two main mechanisms: twinning and dislocation slip [28]. The operative mechanism is affected by the processing conditions and the inherent structure of the deformed alloy during the plastic deformation process [28]. With the addition of Al, the NAB alloys demonstrate a greater tendency to form nanotwins owing to their low stacking fault energy (SFE) [26]. The peculiar martensite phase contains disordered 3R structures with high symmetry [37] and long period stacking orders [38]. However, a high density of stacking faults is also observed in Fig. 17, Fig. 18. For 3R martensite structures that contain complicated periodic stacking faults in their long-range structure, there is a high density of stacking faults [39]. This demonstrated that the twins may nucleate at the grain boundary and thus growing toward the interior of the grain by emissions of partial dislocations at the grain boundary [40]. Molecular dynamic simulations have demonstrated that the formation of a stacking fault is followed by a twin nucleation [41]; when partial dislocations have the same Burgers vector and act as initial dominant partial dislocations, twins can be emitted on adjacent planes to form nucleation twins [26].

As shown in Fig. 18(a), the martensite twins are surrounded by high densities of dislocations at the grain boundaries, indicating that martensite twins may nucleate at the grain boundary and thus growing toward the interior of the grains. Partial dislocations in the grain interior indicate that emissions of partial dislocations from the grain boundary contributed to the growth of martensite nanotwins. Fig. 18(a) also shows that the accumulation of dislocations is concentrated around the fine κ IV phases, indicating that the dislocations motivated the nucleation and growth of the fine κ phases. In this way, the fine and well-distributed κ IV phases make a significant contribution to the increase in hardness. Martensite twins were not found in the UVAM3600 and UVAM4800 samples. Thus, dislocations gradually accumulated at the grain boundaries as the cutting speed increased to 150 m/min, accelerating the formation of nanotwins owing to the low SFE of the NAB alloy.

3.4.4. Contributions of strengthening mechanisms

The yield stress in bulk metals and alloys has been proven to have a positive proportional correlation with hardness [42, 43, 44]. The strengthening mechanisms can be estimated through well-established models and experiment data through calculating the contributions for the yield stress from grain refinement, precipitate strengthening, and dislocation strengthening. The relative contributions to the yield stress have reference significance for the strengthening mechanisms of hardness.

The Hall–Petch model [32,45,46] demonstrates the promotion of grain refinement strengthening.

\[
\Delta \sigma_{HP} = \sigma_0 + K_{HP}d^{-\frac{1}{2}}
\]  

where \( \sigma_0 \) represents the yield stress for the single-crystal; KHP is the Hall–Petch coefficient of the Cu alloy, and \( d \) is the grain size. The refinement of grains produces a shortened grain boundary length, enhancing the resistance to dislocations. The values of \( \sigma_0 \)and KHP are 36.2 MPa and 0.17 MPa·m⁻¹/₂, respectively. In UVAM6000, the size of the matrix was approximately 6 μm. As a result, \( \Delta \sigma_{HP} \) was 106 MPa.

The Orowan equation [32,47,48] describes the precipitate-strengthening effects of the fine κIII phases:

\[
\Delta \sigma_{Or} = \frac{MGb^2}{2\pi \bar{\lambda}} \sqrt{1 - \nu}
\]  

In this model, \( G \) and \( M \) are the shear modulus (42.1 GPa) and the Taylor factor (3.06), respectively; \( r \) is the mean grain size of the κIII phase; \( b \) represents the value of the Burger vector (0.256 nm); \( \nu \) is the Poisson’s ratio (0.34) for α(Cu); \( \bar{\lambda} \) is expressed as

\[
\bar{\lambda} = \frac{4r(1-f)}{f}
\]  

where \( f \) is the volume fraction of fine κIII phases (18%); thus, \( \Delta \sigma_{Or} \) is 68.7 MPa.
The density of dislocations $\rho$ in UVAM6000 can be calculated from the XRD patterns using the Williamson–Hall model [49,50]:

$$B \cos \theta = \frac{K \lambda}{d} + \varepsilon \sin \theta_b$$  \hspace{1cm} (8)

$$\rho = \frac{2\sqrt{3}\varepsilon}{db}$$  \hspace{1cm} (9)

where $\varepsilon$ is the micro strain; $b$ represents the Burgers vector (0.256 nm); $\lambda$ is the wavelength of X-rays; $\theta_b$ is the Bragg angle; $K$ is approximately 0.9. From Eqs. (8), (9), the density of dislocations $\rho$ is $9.74 \times 10^{13}$ m$^{-2}$.

$$\Delta \sigma = M \sigma G b \rho^{\frac{1}{2}}$$  \hspace{1cm} (10)

where $G$ and $b$ are the shear modulus (42.1 GPa) and the value of the Burger vector (0.256 nm), respectively; $M$ and $a$ are the Taylor factor (3.06) and the constant for the Cu-based alloy (0.2), respectively; $\Delta \sigma_\mathrm{d}$ is 64.6 MPa.

For UVAM6000, grain refinement ($\Delta \sigma_{\mathrm{HP}}$) has the greatest contribution, followed by precipitate ($\Delta \sigma_{\mathrm{Or}}$), dislocation ($\Delta \sigma_{\mathrm{d}}$), and nanotwin strengthening. Because the UVAM3600, UVAM4800, and UVAM6000 samples share a similar machining process, grain refinement has the greatest contribution, followed by precipitate and dislocation strengthening for the UVAM3600 and UVAM6000 samples.

4. Conclusion

1) An increase in the cutting speed results in a low friction torque and high wear resistance, which promote the fretting performance of UVAM-processed NAB alloys. Compared to UVAM3600, the Tsta of UVAM6000 was decreased by 29.5%. In addition, the ET of UVAM3600 was 26.6% greater than that of UVAM6000, and the worn surface area in UVAM3600 was 5.6% higher than that in UVAM6000.

2) The microhardness of UVAM-processed NAB alloys increases with an increase in the cutting speed, which contributes to the high fretting properties of UVAM-processed NAB alloys. Moreover, the refinement of the $\kappa$ phases and the effects of dislocations and stacking faults help increase the hardness. The density of dislocations was $9.74 \times 10^{13}$ m$^{-2}$ and the size of the $\kappa$III phase was approximately 100 nm in UVAM6000. Grain refinement, precipitation, and dislocation strengthening were the main mechanisms for all samples.

3) Nanotwin strengthening was observed at a cutting speed of 150 m/min. The formation of martensite phases is attributed to severe plastic deformation and the special 3R martensitic structures of the NAB alloys. Martensite nanotwins may nucleate at the grain boundary and grow toward the interior of the grains.

4) From the analysis of well-established models and experimental data, grain refinement has the greatest contribution, followed by precipitate, dislocation, and nanotwin strengthening for UVAM6000. Because the UVAM3600, UVAM4800, and UVAM6000 samples share a similar machining process, grain refinement has the greatest contribution, followed by precipitate and dislocation strengthening for the UVAM3600 and UVAM4800 samples.

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