Influence of milling on the development of stress corrosion cracks in austenitic stainless steel

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ABSTRACT

We have examined the influence of mechanical surface finishing on the development of residual stresses, and on the subsequent formation of stress corrosion cracks, in 316Ti austenitic stainless steel after exposure to boiling magnesium chloride. The surface residual stresses of as-received plate, prior to machining, were found to be biaxial and compressive. However, abrasive grinding produced significant compressive stresses in the machining direction but much lower perpendicular stresses. On the other hand, milling produced high biaxial tensile stresses (approaching the ultimate tensile strength, UTS, of the material), which were found to be relatively insensitive to cut depth but to vary as a function of feed rate. On the milled surfaces a distinctive pattern of stress corrosion cracking was evident with longer primary cracks nucleating along the milling direction and secondary, shorter, cracks nucleating perpendicularly. As the surface tensile stress was lower perpendicular to the milling direction, we postulate that the nucleation of primary cracks parallel to machining must be driven by the surface profile after machining (and associated micro-stresses) as much as by the macroscopic residual stresses.

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1. Introduction

Environmentally assisted cracking is one of the most harmful localised damage processes and encompasses a wide range of mechanisms that includes, for example: hydrogen induced cracking, hydrogen embrittlement, corrosion fatigue and stress corrosion cracking. Typically initiating at local physical features in the material, cracks subsequently develop through various stages of growth from: (i) the formation of multiple short cracks, (ii) the coalescence of these cracks and, eventually, (iii) generation of a dominant long crack that propagates to failure. Cracking generally starts at local defects, which may be microstructural features within the body of the material or, more typically, commence from surface features that are initially present as a consequence of materials processing (e.g. local microstructure, surface roughness) or arise from an in-service damage process such as wear, erosion, or corrosion (e.g. pitting). For austenitic stainless steels, which are the first-choice workhorse alloys for industrial applications requiring corrosion resistance, a key susceptibility is to stress corrosion cracking in environments containing chloride ions where an applied (i.e. service) or residual tensile stress is also present.

Chloride-induced stress corrosion cracking of ferrous alloys necessarily commences from a component surface because access to the external environment is required and generally occurs on austenitic microstructures, since ferritic phases are relatively immune from such damage. Thus, the nature of the material surface and near sub-surface (i.e. microstructure, near-surface residual stress and surface geometry) is critical to the initiation and propagation of stress corrosion cracks.

Machining involves considerable localised plastic deformation, generating thermal energy, both of which might give rise to residual stresses. The controlling factors for the generation of surface residual stress during machining were first evaluated by Henriksen (1951). He suggested (for low carbon steel) that residual stresses are primarily generated from plastic deformation rather than differential thermal expansion. For carbon steels, the nature of the residual stress is somewhat dependent upon the hardness of the materials, thus Matsumoto et al. (1986) showed that stresses are tensile for softer steels and tend to become compressive for harder steels. However, in most cases a tensile stress state is left at the surface after machining. According to Brinkmeier (1987) the size of this tensile residual stress, and the depth of the region influenced by the stress, tend to increase with feed rate and cutting speed. For pure turning operations, Leskovar and Peklenik (1981) showed that tensile residual stresses dominate and increase with turning speed; similarly El-Khabeery and Fattouh (1989)
showed that for pure milling operations a greater depth of cut and greater feed rates led to larger and deeper tensile residual stresses.

Regarding austenitic steels, Boothroyd (1975) has noted that they are particularly challenging to machine because of their high work-hardening rate and their galling tendency. Jang et al. (1996) and M’Saoubi et al. (1999) found that, after turning of cylindrical specimens, a plane stress condition existed on the machined surface where the hoop stresses were predominantly tensile (and influenced by the cutting conditions), with the axial stresses compressive (and relatively independent of cutting conditions). Kuroda and Marrow (2008 a,b) examined a range of turning conditions in austenitic stainless steel, also finding that the hoop stress was tensile, while the axial stress could be tensile or compressive; tensile stresses tended to develop with high feed rate and low cut depth. The sub-surface hoop and axial stresses were compressive, and insensitive to cutting conditions. In conventional cutting the volume of the thermally affected zone is relatively small compared with the zone of plastic deformation. However, the given the low thermal conductivity of austenitic steels, computer simulations also by Jang et al. indicated that thermal expansion of the surface during machining would occur resulting in greater tensile stresses on cooling compared with ferritic materials. Peyre et al. (2000a,b) have demonstrated changes in the surface microstructure of austenitic alloys after peening with influence on corrosion resistance. This was ascribed to the intrinsic high work-hardening rate of austenite combined with the rapid surface deformation that resulted in a high near-surface dislocation density and it seems likely that this mechanism would also be valid during high speed machining. On the other hand, Miguelez et al. (2009) suggested that residual stresses arise from thin thermally-affected layers which produce thermal expansion and subsequent plastic flow, though such a mechanism may also be dominated by residual stresses generated at the interface at the base of the sheared chip of the plastic region with the surrounding elastic material. Overall, previous work confirms that the surface residual stress distributions in alloys after final surface machining depends on a number of complex and interrelated parameters, including: cutting speed, feed rate, depth of cut and tool geometry as well as the nature of the near-surface microstructure.

Many mechanical failures in service result from an interaction of stresses in the material and the environment. One critical process is stress corrosion cracking (SCC) that, for austenitic stainless steels, almost always initiates from a pre-existing corrosion pit and is largely controlled by the chloride ion concentration, temperature and time in service. Surface preparation plays a significant role in aiding corrosion pit nucleation through the combined or independent effects of: (i) geometry associated with surface roughening, and (ii) the influence of roughness on surface chemistry. For example, it is well known that the localised corrosion susceptibility of stainless steel in chloride solutions is significantly affected by surface finish. Thus, Burstein and Pistorius (1995) and Zuo et al. (2002) both found that metastable pits initiated more easily on rougher surfaces because of the greater number of sites available for such pitting to occur. However, metastable pits have a higher probability of transforming to stable pits on smooth surfaces since more rapid diffusion rates tends to prevent re-passivation. This is because, in many cases, the survival of pit precursors (i.e. metastable pits) has been shown to depend on the maintenance of an effective diffusion barrier formed by salt films (Hong and Nagumo, 1997) or by lacy metal covers over pit mouths (Ernst and Newman, 2002). Additionally, Moayed et al. (2003) qualitatively demonstrated, using both potentiostatic and potentiodynamic critical pitting temperature experiments, that the pitting resistance tends to increase with increasing surface roughness.

Stress corrosion cracking occurs as a chemo-mechanical embrittlement phenomenon in nominally tough and ductile alloys at stress intensity factors (KIC) considerably lower than the nominal fracture toughness of the material. For example Vinoy et al. (1996) found that for AISI316L steel in acidified boiling sodium chloride, the critical threshold for the development of SCC KSCC was 13 MPa m^-1/2 (for annealed material) and 10.5 MPa m^-1/2 (for sensitised material) and for austenitic stainless steels KSCC is generally between 10 and 20 MPa m^-1/2. Recently, there has been interest in studying the effect of residual fabrication stresses, primarily cold work, on the susceptibility of stainless steels to SCC, particularly in high temperature water for nuclear applications (Tice et al., 2009) but also in chloride environments (Ghosh et al., 2011) where they found that cold working resulted in significant local formation of deformation-induced martensite in AISI304 and hence increasing susceptibility to SCC. The influence of surface finish (Ra) on SCC of AISI304 under simulated atmospheric corrosion conditions, as a consequence of differing surface finishing operations (predominantly grinding and abrasive wheel milling), was found to result in very high levels of surface tensile residual stress (~1000 MPa, determined by hole drilling) with stress corrosion cracks on ground surfaces found to originate at corrosion pit sites (Turnbull et al., 2011). The influence of surface microstructure on SCC of machined AISI304 stainless steel in a “U”-bend geometry was studied by Ghosh and Kain (2010) where they ascribed a five times increase in the crack density (i.e. number of cracks per unit surface area) in machined samples compared with annealed samples to the surface tensile stresses (which were not measured), surface grain refinement and surface martensite formation. However, many of these studies are unsatisfactory in neither quantifying crack morphology nor surface stresses nor surface roughness.

Overall, therefore, the effect of surface roughness and sub-surface residual stress on the initiation and propagation of stress corrosion cracks remains unclear. The aim of this work, therefore, is to examine how stress corrosion cracking develops as a function of varying machining parameters and, hence, whether there is a systematic relationship in observed cracking between local surface morphology (i.e. machining profile) and residual stresses present in the material.

### 2. Experimental procedure

#### 2.1. Sample preparation

Rectangular samples (10 × 12 × 19 mm) were cut from a rolled plate of AISI 316Ti (16.9%Cr, 10.8%Ni, 2.0%Mn, 1.6%Mn, 0.59%Ti, 0.08%C, balance Fe). This alloy is a stabilised grade of austenitic stainless steel that is used in moderately elevated temperature applications where the preferential precipitation of titanium carbide rather than chromium carbide provides resistance both to sensitisation (i.e. grain boundary chromium depletion) and to creep by grain boundary pinning. The physical and mechanical properties of the alloy are representative of the AISI 300 family of austenitic steels. However, the room temperature yield stress and high temperature creep resistance are somewhat larger (compared with 316L) as a result of the presence of the titanium carbide precipitates, typically around 1 μm in dimension, that tend to nucleate on grain boundaries. The flat surfaces (i.e. the longitudinal-transverse, L-T, direction) from the original plate were left as-received (i.e. mill finish). The two plate ends (i.e. the short-transverse, S-T, direction) were coarse-cut using a bandsaw, while one of the plate edges (i.e. the longitudinal-short, L-S, direction) was ground parallel to the rolling direction to provide a nominally flat surface (Fig. 1).

The remaining plate edge was machined using a Hurco Hawk 30 milling machine, using a Sandvik Coromill general-purpose, solid
carbide, 4-flute, endmill, 5 mm in radius, at a tangential velocity of 1.667 m s\(^{-1}\). Copious coolant was used to minimise the build-up of heat during machining and a new tool was used for each sample. This allowed three surface finishes (i.e. as-received, ground and milled) to be examined on a single sample. The combinations of milling parameters, at constant tangential velocity, are given in Table 1 and were selected using the procedure of Kuroda and Marrow (2008a,b, Journal of Materials Processing Technology). This methodology is based on the Tamaguchi approach where combinations of milling parameters recommended for 300 series stainless steel were selected.

### 2.2. Stress corrosion crack generation

Stress corrosion cracks were induced in the samples by immersion in boiling magnesium chloride according to ASTM-G36 (1994) and are reported as a function of the milling parameters. Samples were exposed in a round-bottomed flask containing provision for a thermometer and a water-cooled condenser. Technical grade magnesium chloride hexahydrate was held at a temperature of 155 ± 5 °C for a period of two weeks by means of a heating mantel connected to a temperature controller. Sufficient deionised water was added at the beginning of the experiment to promote dissolution at the operating temperature. Although the condenser reduces the amount of vapour lost during boiling, deionised water was added daily to replenish the small losses that occurred. The samples were held upright within the flask using a PTFE tray of approximately 5 mm thick and 95 mm in diameter that was designed to prevent any direct contact between the samples and the glass; the corrosive solution was free to flow freely between the samples. Sample roughness was found to be unchanged after this treatment.

### 2.3. Sample characterisation

The general pattern of surface cracking and corrosion was observed using a Leitz Metallolvert light microscope. In addition, a Bruker Contour GT-K1 white light interferometer comprising an optical metrology module was used for the acquisition of high vertical resolution optical images of the milled surfaces. Acquired images were processed to remove noise with a low pass filter applied via a fast Fourier transform; this enabled the milling profiles to be better resolved from the intrinsic roughness of the surfaces. These three-dimensional surface scans were used to measure roughness values, \(R_s\), to measure crack lengths and also to determine whether cracks were located at peaks or troughs of the surface profile.

The samples were initially characterised by mechanical testing where the results of duplicate tests fall within 1% of each other. The 0.2% proof stress of the as-received alloy was 235 MPa while the true strain at failure was 38% at a true ultimate tensile stress (UTS) of 870 MPa. Optical metallography was carried out in both the short-transverse (\(S-T\)) and the long transverse (\(L-T\)) directions. The average grain size of this material was determined using the linear intercept method to be 20–30 μm in the transverse direction and 40–50 μm in the longitudinal (rolling) direction. Stringers of retained delta ferrite, aligned along the rolling direction, were evident in the as-received material.

### 2.4. Residual stress measurement

A Proto iXRD, which is a diffractometer that has an X–Y position-controlled X-ray head, was used for the measurement of residual strain data before and after exposure of the samples in boiling magnesium chloride. This equipment uses the \(\sin^2 \psi\) method whereby strains are calculated by fitting the top 75% of the peak intensity of the selected diffraction peak, (\(311\)) in this case, to standard Gaussian profiles. The d-spacing is then calculated from the peak position at each \(\psi\) angle according to Bragg’s law, and the slope of the d-spacing versus \(\sin^2 \psi\) plots are used to calculate the strain, which was converted to stress using a Young’s modulus of 195 GPa and a Poisson’s ratio of 0.29 (Peckner and Bernstein, 1977), values that are representative of austenitic steels.

A manganese X-ray source (\(\lambda = 0.210 \text{nm}\)) with a 2 mm collimator were used such that between forty and one hundred grains were sampled during each diffraction measurement within a mean penetration depth of around 6–7 μm. The data comprised three sets, each containing nine successive measurement points along the same line, with the points spaced at 2 mm apart along lines that were at 3, 6, and 9 mm from the same edge; these data were averaged (27 points per measurement) and are reported below for orientations parallel and perpendicular to the rolling direction.

### Table 1

<table>
<thead>
<tr>
<th>Sample identification</th>
<th>Milling parameters (4-flute end milling tool at a tangential velocity of 1.677 m s(^{-1}))</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Depth of cut (mm)</td>
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<tr>
<td>1</td>
<td>0.4</td>
</tr>
<tr>
<td>2</td>
<td>1.2</td>
</tr>
<tr>
<td>3</td>
<td>1.4</td>
</tr>
<tr>
<td>4</td>
<td>0.7</td>
</tr>
<tr>
<td>5</td>
<td>0.6</td>
</tr>
<tr>
<td>6</td>
<td>1.9</td>
</tr>
</tbody>
</table>

### Table 2

<table>
<thead>
<tr>
<th>Sample number</th>
<th>As-received surface</th>
<th>Ground surface</th>
<th>Milled surface</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>0° (MPa)</td>
<td>90° (MPa)</td>
<td>0° (MPa)</td>
</tr>
<tr>
<td>1</td>
<td>–270 ± 30</td>
<td>–250 ± 20</td>
<td>–250 ± 40</td>
</tr>
<tr>
<td>3</td>
<td>–275 ± 40</td>
<td>–250 ± 40</td>
<td>–310 ± 50</td>
</tr>
<tr>
<td>4</td>
<td>–240 ± 50</td>
<td>–220 ± 45</td>
<td>–300 ± 30</td>
</tr>
<tr>
<td>5</td>
<td>–260 ± 35</td>
<td>–260 ± 25</td>
<td>–400 ± 30</td>
</tr>
<tr>
<td>6</td>
<td>–250 ± 25</td>
<td>–220 ± 35</td>
<td>–300 ± 20</td>
</tr>
</tbody>
</table>
Fig. 2. (a) As-received sample with cracks propagating from the base of shallow surface pits (parallel to the short-transverse direction). (b) As received sample showing crack deflection due to the presence of delta ferrite stringers (parallel to the longitudinal-transverse direction).

Fig. 3. Typical network of cracks formed on the milled surface with feed rate 0.025 mm/tooth and cut depth of 0.4 mm (sample no. 1).

3. Results

3.1. Residual stress

The residual stress within the as-received samples (i.e. prior to machining) was biaxial and compressive ($-250 \pm 30$ MPa) both parallel and perpendicular to the principle rolling direction of the plate, Table 2. After grinding, the surface stresses remain compressive, but have become generally uniaxial in the grinding (longitudinal) direction with lower stresses in the short-transverse direction. On milled surfaces the residual stresses were all tensile and biaxial with many values exceeding the uniaxial proof stress of the as-received material ($\sim 235$ MPa) and approaching the uniaxial UTS ($\sim 870$ MPa).

Fig. 4. Primary and secondary cracks on milled surfaces with local crack coalescence evident: (a) sample no. 2 (b) sample no. 3, (c) sample no. 4, and (d) sample no. 6.
Table 3
Primary crack length and frequency as a function of residual stress and roughness for each milled profile (sample 5 was not analysed).

<table>
<thead>
<tr>
<th>Sample number</th>
<th>0° (MPa)</th>
<th>90° (MPa)</th>
<th>Average $R_a$</th>
<th>Depth of tensile sub-layer (μm)</th>
<th>Primary crack lengths (μm)</th>
<th>Secondary cracks (mm$^{-1}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>390 ± 120</td>
<td>240 ± 90</td>
<td>0.87</td>
<td>20–30</td>
<td>&lt;150</td>
<td>17</td>
</tr>
<tr>
<td>2</td>
<td>780 ± 220</td>
<td>680 ± 150</td>
<td>0.32</td>
<td>40–50</td>
<td>&gt;200</td>
<td>24</td>
</tr>
<tr>
<td>3</td>
<td>420 ± 40</td>
<td>290 ± 90</td>
<td>0.51</td>
<td>30–40</td>
<td>&lt;250</td>
<td>14</td>
</tr>
<tr>
<td>4</td>
<td>620 ± 40</td>
<td>380 ± 40</td>
<td>0.39</td>
<td>20–30</td>
<td>&gt;450</td>
<td>11</td>
</tr>
<tr>
<td>6</td>
<td>760 ± 40</td>
<td>690 ± 75</td>
<td>0.34</td>
<td>30–40</td>
<td>&lt;150</td>
<td>11</td>
</tr>
</tbody>
</table>

3.2. Pitting and cracking morphology

On the as-received surfaces, where biaxial compressive residual stresses were present (in the longitudinal and the longitudinal-transverse directions), and on ground samples, where a nominally uniaxial compressive residual stress was present in the longitudinal (grinding) direction, small pits were evident approximately 2–5 μm in size. Where cracks formed, they always propagated from the base of these pits, Fig. 2(a), however not all pits were associated with cracks. Where propagating stress corrosion cracks intersected with delta ferrite stringers (which are resistant to chloride-induced stress corrosion cracking) then they tended to be deflected along the rolling direction, Fig. 2(b).

Significantly on the flat milled surfaces, where biaxial tensile residual stresses were present, the corrosion morphology was completely different from the other surfaces and extensive damage was evident in the form of multiple and interacting cracks. A typical example of this damage is shown in Fig. 3 from which the cracking can be seen to form a regular pattern; significantly, this is not associated with pre-existing pits but clearly has some relationship with the machining marks. Additional examples for other samples are shown in Fig. 4.

Generally primary cracks appear to be associated with the asperities due to milling, while secondary cracks, all generally of similar length (~25 μm), branch orthogonally from the primary cracks. Fig. 5 shows 3-D isometric projections of selected surfaces that illustrate several common features of the stress corrosion cracking process: (a) primary cracks tend to nucleate within the troughs would influence the stress corrosion cracking behaviour. However, over the range of cut depths and feed rates used here, the arithmetic average surface roughness ($R_a$) does not appear to be a useful measure as it does not vary in any consistent manner, falling in the range 0.87 μm (0.4 mm cut depth and 0.025 mm/tooth feed rate) to 0.32 μm (1.2 mm cut depth and 0.04 mm/tooth feed rate), although for a wider range of conditions, Kuroda and Marrow (2008a,b, Journal of Materials Processing Technology) deduced a relationship for surface roughness with both feed rate and cut depth using a statistical fit. Nevertheless microscopy confirms that there is periodicity in the surface profile (i.e. troughs and peaks) as is usually, not at pits. This

4. Discussion

4.1. Residual stress distribution

For milled samples the tensile residual stress is biaxial with somewhat greater tensile stresses being generated in the direction of machining. The surface residual stresses were found to be relatively insensitive to cut depth while they increase substantially with feed rate nominally approaching the ultimate tensile strength of the alloy at a feed rate of 0.08 mm/tooth (cut depth of 0.6 mm). The depth of the tensile sub-layer on each sample was estimated from the depth of cracking and lay in the range 20–50 μm. For the sharp tools used in this work, feed rate might be expected to have a much greater influence on energy input (plastic work) into the alloy compared with the cut depth. However, as pointed out by Basuray et al. (1977) and many others, blunt tools require much larger machining forces, consequently the effect of cut depth is expected to be considerably greater where the tool is worn.

4.2. Surface profile

The surface roughness was measured on the premise that a particular machining operation should result in consistent changes to the surface profile of the material and that the overall roughness would influence the stress corrosion cracking behaviour. However, over the range of cut depths and feed rates used here, the arithmetic average surface roughness ($R_a$) does not appear to be a useful measure as it does not vary in any consistent manner, falling in the range 0.87 μm (0.4 mm cut depth and 0.025 mm/tooth feed rate) to 0.32 μm (1.2 mm cut depth and 0.04 mm/tooth feed rate), although for a wider range of conditions, Kuroda and Marrow (2008a,b, Journal of Materials Processing Technology) deduced a relationship for surface roughness with both feed rate and cut depth using a statistical fit. Nevertheless microscopy confirms that there is periodicity in the surface profile (i.e. troughs and peaks) that is caused by the milling process. While there was no obvious relationship between pit initiation and location (i.e. surface profile), stress corrosion cracking in the milled surfaces almost always initiated in the troughs between the peak asperities and, unusually, not at pits.

![Fig. 5. Isometric projections of milled surfaces showing typical stress corrosion cracks: (a) sample no. 2 (secondary cracks have grown between adjacent milling peaks); (b) sample no. 3 (primary cracks nucleated between milling marks within troughs, or on the shoulder, of asperities.)](image-url)
is presumably a function of local stress distribution however the methodology used in this work is not able to resolve micro-strain at the requisite scale length (of the order of 5–10 μm) in order to confirm this hypothesis.

4.3. Stress corrosion cracking

The pattern of cracking in the milled samples is perhaps the most interesting and significant observation in this research. Stress corrosion cracks would tend to initiate perpendicular to the dominant stress in mode I (crack opening) loading. Here we measure somewhat larger tensile residual stresses parallel to the milling direction with lower stresses perpendicular to this direction. Consequently the longer, primary, cracks might be expected to develop perpendicular to the milling direction, however, it is the shorter, secondary cracks, that are evident in this direction with the primary cracks developing along the geometric milling marks. This strongly suggests that the local geometry is at least as influential in the development of the stress corrosion cracks as the dominant direction of tensile stress. The majority of the secondary cracks have arrested at approximately the mid-point between the primary cracks. This might be presumed either to be a consequence of local compressive stresses between the milling lines that are below the resolution of the measurement or, perhaps more likely, an outcome of local relief of tensile stress after the cracking process has commenced. However, this would require a detailed analysis using finite element modelling of local stresses in the presence of multiple cracking and is beyond the scope of this paper.

5. Conclusions

(1) Surface residual stresses have been mapped in milled austenitic stainless steel as a function of tool feed rate and cut depth. We find that flat milled samples have a biaxial tensile surface stress with a higher stress in the direction of milling.

(2) Primary stress corrosion cracks were observed in milled samples where the cracks were aligned with the milling marks. Unusually, such cracks were perpendicular to the lower surface tensile stress direction and demonstrate that surface morphology is as influential on crack path as tensile stress.

(3) Extensive secondary cracking was observed orthogonal to the primary cracks. Such cracks were perpendicular to the higher of the surface tensile stress directions but arrested half-way between the primary cracks. This presumed to be due to local stress-relief during cracking.

References


