Influence of microstructure on fatigue crack nucleation and microstructurally short crack growth of an austenitic stainless steel

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\begin{abstract}

In this study, the effect of microstructure on crack nucleation and microstructurally short fatigue crack growth is investigated for a metastable austenitic stainless steel. Fatigue tests were conducted at an intermediate fatigue life regime between $10^4$ and $10^6$ cycles such that martensitic phase transformation occurs given sufficient localized deformation. Through the use of scanning electron microscopy, along with electron backscatter diffraction, several micro-cracks were analyzed and compared. The influence of microstructural features such as twin boundaries, slip band intrusions/extrusions, grain boundaries, inclusions, and martensitic transformed areas on the crack initiation life is discussed. The initiation stages of crack nucleation and the subsequent microstructurally short crack growth for each microstructural feature are compared revealing that twin boundaries and slip bands are the most dominant initiation features. However, the initiation mechanism governing crack nucleation for each feature was different. Additionally, the phase transformation behavior showed only minor effects on the microstructurally short crack growth leading up to an engineering crack. It was found that while the cracks that propagated more quickly had larger transformed martensitic zones around the crack tip, this was due mostly to the size of the crack. Interestingly, the initiation life in the transitional fatigue regime was observed to be more sensitive to crack initiation feature than the martensitic transformation.

\end{abstract}

\section{Introduction}

Traditionally, fatigue life is separated into the initiation stage and the fatigue crack propagation stage. The crack initiation stage can be further separated into the life to nucleate a micro crack and the subsequent microstructurally short crack (MSC) growth that continues until the crack is long enough not to be directly influenced by the microstructure. The crack initiation is heavily affected by the microstructural features, and thus, there is a need to fully understand the microstructure-fatigue property relationships. It is well established that cracks initiate at a variety of microstructural features such as twin boundaries (TB), grain boundaries (GB), slip band extrusion/intrusions (SB), and inclusions (IN) in many austenitic stainless steels \cite{1–6}. Previous research studies have shown that crack initiation in the high cycle fatigue regime (fatigue lives of $N_f > 10^6$) is dominated by twin boundaries rather than other microstructural features \cite{1–5}, while crack initiation during low cycle fatigue (fatigue lives of $N_f < 10^6$) is dominated by inclusions for many austenitic stainless steels \cite{6}. This is in contrast to most metals which typically show crack initiation at inclusions in the high cycle regime and other microstructural features such as slip bands and grain boundaries in the low cycle regime. Additionally, the majority of the high cycle fatigue life is spent initiating an engineering crack \cite{7,8}, defined as a fatigue crack with a length on the order of a few average grain diameters. In contrast, the majority of the low cycle fatigue life is spent in the crack propagation stage where plastic deformation results in much quicker crack initiation \cite{7}. The transitional fatigue regime between the low and high cycle regime ($10^4 < N_f < 10^6$) would then be expected to show a mixture of low and high cycle crack initiation behavior since the ratio of plastic and elastic strain is near unity.

For metastable austenitic stainless steels such as 304L (304LSS), the effect of microstructure on crack initiation can be complicated by the austenite, $\gamma$ (FCC), to martensite, $\alpha'$ (BCC), phase transformation ($\gamma \rightarrow \alpha'$ phase transformation) \cite{9–15}. During high cycle fatigue, the plastic strain is not typically large enough to show significant initiation of the $\gamma \rightarrow \alpha'$ transformation \cite{16}. As the load approaches the transitional fatigue regime, however, the increase in plastic strain may result in martensitic transformation. The effect of this phase transformation leads to cyclic hardening of the material from a combination of (1) the higher strength of the $\alpha'$ phase, (2) the increased dislocation density

\begin{keyword}

Crack initiation \\
Microstructurally small crack growth \\
Martensitic transformation \\
Microstructure \\
304L stainless steel \\
Fatigue

\end{keyword}
from the $\gamma \rightarrow \alpha'$ phase transformation [17,18], and (3) the subsequent interaction of these dislocations. The transformation from austenite to martensite also results in an increase in volume of approximately 2% [13], which can affect the growth behavior of MSC due to transformation induced crack closure [19]. Pineau and Pelloux [10] showed that fatigue crack growth rates were higher in a martensitic stainless steel when compared to stable austenitic stainless steels. It remains, however, to understand how the $\gamma \rightarrow \alpha'$ phase transformation during cyclic loading affects the MSC growth behavior for nucleated cracks and the life to initiate a dominant crack in the material.

The primary purposes of this investigation are to determine the effect of microstructure on fatigue crack initiation and to study the development of martensite in the vicinity of microstructurally short cracks (MSC) in a commercially available 304L stainless steel. Additionally, the effects of microstructure and the $\gamma \rightarrow \alpha'$ transformation on the crack initiation stage are investigated in a transitional fatigue life regime (i.e. $\frac{A_0}{2} \approx 1$).

2. Material and experimental program

2.1. Material and specimen design

The 304LSS used in this study was delivered in cold drawn and annealed %" diameter bars with the chemical composition listed in Table 1 and the reported mechanical properties listed in Table 2. Two specimen types were used for this study, as shown in Fig. 1. Traditional cylindrical fatigue specimens with uniform gage section as indicated in Fig. 1(a) were machined longitudinally from the as-received bar. These specimens were carefully ground smooth using a series of 500, 800, and 1200 grit silicon carbide papers. Round cornered square gaged fatigue specimens were carefully ground smooth using a series of 500, 800, and 1200 grit silicon carbide papers. Before testing, SGF specimens were electrochemically polished using a Struers Lescropol 5 with a perchloric acid electrolyte (A2, Struers) at 45 V for 40 s and not allowed to exceed 32 °C. The electro-polishing allowed high quality diffraction patterns to be obtained using EBSD even after substantial cyclic loading. The EBSD maps were collected using EDAX-TSL software at an acceleration voltage of 20 kV with step sizes ranging from 0.1 to 0.5 µm using a hexagonal scan grid.

The as received microstructure consisted of equiaxed grains with a high density of annealing twins. Fig. 2 details the as-received microstructure for the cross-section of a selected SGF specimen after electro-polishing and before testing. Fig. 2(a) is an inverse pole figure map (IPF) from electron backscatter diffraction (EBSD) showing equiaxed grains with a high density of annealing twins. Fig. 2(b) presents the grain size distribution for the given IPF map showing an average grain size around 54 µm for the scanned area. The average grain size was verified using a grain boundary etch and lineal intercept method. Fig. 2(c) gives a representation of the martensite phase distribution throughout the microstructure and shows a fully austenitic microstructure. While the microstructure was shown to be fully austenitic, there were some trace amounts of $\delta$-ferrite. These $\delta$-ferrite were elongated in the direction of the cold drawing and were removed by the electro-polish, which resulted in elongated voids in the longitudinal direction of the SGF fatigue specimens. Fig. 2(d) presents the fraction of twin boundaries (blue) for the selected area. Annealing twins are known to have a specific misorientation of 60° which have an allowable deviation of 8.66° from the ideal $\Sigma = 3$ boundary according to the Brandon criterion. Thus, twin boundaries were defined as the ones with $60° \pm 8.66°$ misorientation, while grain boundaries were defined as all other boundaries above 5° misorientation.

2.2. Fatigue testing

A series of interrupted and non-interrupted fully reversed uniaxial load controlled fatigue tests were conducted on a closed loop servo-hydraulic MTS 810 fatigue testing frame with a load capacity of 100 kN. Non-interrupted tests using the specimen design shown in Fig. 1(a) were conducted at stress amplitudes of 300, 330, 350, and 400 MPa with sinusoidal wave forms in order to construct a stress-life ($\alpha_n - 2N_c$) curve. An extensometer was used to record the cyclic strain and the generated stress-strain behavior was used to determine a stress level that was within the transitional fatigue regime. Due to a limited number of samples from the same production batch of material, only two tests were conducted for each stress amplitude except for the runout stress amplitude, for which only one test was conducted.

Previous research has shown that for a Fe-18Cr-6.5Ni-0.19C stainless steel, plastic strain amplitudes greater than 0.18% are capable of...
generating significant amounts of the α’ phase [10] while another study reported the α’ phase transformation to occur at total strain amplitudes as low as 0.2% [14]. The cyclic stress and strain data showed that a stress amplitude of 330 MPa produced a plastic strain amplitude of 0.18% and a plastic to elastic strain ratio of approximately 0.9 at half life. This ratio was determined to be in the desired transitional region of the fatigue life.

Interrupted tests using the SGF specimens shown in Fig. 1(b) were then performed at a stress amplitude of 330 MPa, which approximately corresponded to the transition from the low to the high cycle fatigue regime. Tests were interrupted every 10,000 cycles until multiple micro-cracks were observed on each surface using scanning electron microscopy (SEM) after which tests were interrupted every 3000–5000 cycles. MSC growth rates (da/dN) were calculated as (2ai−2ai−1)/(Ni−Ni−1) where i is the current cycle count. Due to the tortuosity of the MSC cracks, the crack lengths (2a) were calculated by summing the straight line lengths of the surface cracks from tip to tip. During the interrupted period, the specimen was removed from the testing equipment and carefully aligned in the SEM stage as described in the following section. Interrupted testing was then continued until multiple cracks of approximately 200 µm were observed. At this length, the crack initiation stage (i.e. the combination of crack nucleation and microstructurally short crack growth stages) was assumed to be completed with only the long crack propagation stage remaining.

### 2.3. SEM/EBSD data collection

Scanning electron microscopy was used to detect micro-cracks between 20 and 50 µm on the surface of the SGF specimens during the interrupted fatigue testing. Once a crack was discovered, a secondary electron image was taken of the surrounding surface to record the position of each crack on the surface. During subsequent interruption periods, each crack was located and new cracks were detected.

After imaging, the specimen was tilted so that the surface was at an angle of 70° from the electron beam in order to capture EBSD maps. Surface grain orientation and phase distribution maps were collected along with Taylor and Schmid factor maps for the γ phase by applying the loading direction and the primary FCC slip system of (111) planes in the <110> direction to the grain orientation datasets. Previous research has shown that a relatively large mismatch of Schmid factors of adjacent grains can increase the localized stress across the given boundary which can result in crack initiation along the boundary [20]. Additionally, Taylor factor mismatches (TFM) have been suggested to be an important factor for twin boundary cracking. A large TFM across a twin boundary can be indicative of a strain incompatibility resulting in twin boundary cracking [3,14,21]. For grain and twin boundary cracks, the Schmid factor mismatch (SFM) and Taylor factor mismatch (TFM) values between grains on either side of the boundary were of particular interest as they can be indicative of stress concentrations and elastic incompatibilities across the boundaries. Taylor factors were further classified as either low (2.24–2.68), medium (2.69–3.16), or high (3.17–3.64) with the subsequent mismatches being low-low (LL), low-medium (LM), low-high (LH), medium-medium (MM), medium-high (MH), or high-high (HH) as described by Roach et al. [3].

Crack initiation features were classified as either twin boundary (TB), grain boundary (GB), slip band (SB), inclusion (IN), or martensite (M) cracks. Four SGF specimens were tested under the given conditions and a total of sixty-seven cracks were analyzed to obtain a variety of crack initiation feature types for data analysis.

### 3. Experimental results

#### 3.1. Cyclic behavior

The stress-life plot presented in Fig. 3 along with the SGF specimens data indicate that crack nucleation, MSC growth, and long crack growth stages were all considerable at a stress amplitude of 330 MPa. Under the constant amplitude load control conditions, cyclic softening and cyclic hardening were observed as an increase and decrease in strain amplitude, respectively. Fig. 4 presents the cyclic stress-strain behavior associated with the stress amplitude of 330 MPa. Stage (1) represents the cyclic softening stage in which a repeated stress amplitude results in a larger strain amplitude for the following cycle. This occurs as the microstructure moves into a more stable configuration by an increasing of mobile dislocations [11,22]. Stage (2) represents the cyclic hardening stage in which the same stress results in a lower strain amplitude. At this stage, the microstructure aligns into a more stable configuration which results in the activation of multiple slip systems for a given grain and results in the interaction of the activated slip systems. Additionally, the intersection of the activated slip systems results in γ → α’ phase transformation further contributing to the cyclic hardening observed in the material [11]. Previous research has shown a strong correlation...
between secondary hardening and the volume fraction of the α’ phase [14]. These stages are evident in Fig. 4(a). Fig. 4(b) presents the hysteresis loops for the first cycle as well as cycle 20,000 showing a small amount of tensile cyclic creep which may be attributed to a small amount of pre-strain left over from the drawing process [22].

The interrupted cyclic tests on SGF specimens present an opportunity to observe the changes in microstructure on the surface over the early stage of the fatigue crack initiation. Periodically, when collecting initiation and crack propagation EBSD maps, surface maps were taken over larger areas to observe any microstructural changes on the surface without the influence of the growing crack. The rotation of the grains was readily observable during the interrupted load controlled fatigue tests as indicated by Fig. 5 showing the same area of the microstructure after a) 20,000 cycles and b) 35,000 cycles (loading direction is horizontal in respect to Fig. 5). The superimposed inverse pole figures (IPF) in Fig. 5 shows the rotation for each grain on opposing sides of a twin
favored in a particular direction when some pre-strain is present in the loading axis, other slip systems can become active and deformation may occur on two slip systems. Chai and Laird showed cyclic creep is mainly one active slip system as indicated by the single direction of the strain hysteresis loops for the first cycle and cycle 20,000.

The interaction of multiple slip systems can be very important in the martensitic phase transformation behavior as these intersections often result in the nucleation of the \( \alpha' \) phase within a grain rather than at the grain boundary [24]. This is also evident in Fig. 5 where the \( \alpha' \) phase shows nucleation and growth in grains that are not associated with grain or twin boundaries.

The phase transformation away from the surface was found to be comparable to the phase transformation at the surface in the absence of a growing crack. The parent austenite to martensite transformation showed the typical Kurdjumov-Sachs relationship (i.e. (111) \( \gamma \) \( \|\| \) (110) \( \alpha' \) and \( < 110 > \gamma \|\| < 111 > \alpha' \)). Additionally, the \( \alpha' \) phase in the absence of the \( \varepsilon \) (HCP) phase has been reported to have a blocky irregular shape; however, when it is formed in the presence of the \( \varepsilon \) phase, it becomes more lath like [27]. Only blocky type martensite was found in all EBSD phase maps, indicating the \( \varepsilon \) phase was either non-existent or un-observable in this study. While nucleation of the \( \alpha' \) phase was not captured in this study, nearly all \( \alpha' \) areas were associated with either a twin boundary, grain boundary, or intersecting slip traces under cyclic loading.

4. Crack initiation mechanisms

Fractography provided little evidence into the initiation mechanisms during the fatigue tests on traditional specimens, and therefore, the interrupted test specimens were used to capture crack nucleation and MSC growth information for the stress amplitude of 330 MPa. As previously mentioned, the crack initiation stage includes nucleation and MSC growth leading up to an engineering crack with a length up to a few average grain diameters (roughly 100–200 \( \mu \)m for 304LSS). The crack initiation microstructural feature for each micro-crack was determined using EBSD. The nucleation of a micro-crack at a microstructural feature was determined for twin boundaries, grain boundaries, and MSC growth information for the stress amplitude of 330 MPa. A few average grain diameters (roughly 100–200 \( \mu \)m for 304LSS) was used to capture crack nucleation and MSC growth information for the stress amplitude of 330 MPa. As previously mentioned, the crack initiation stage includes nucleation and MSC growth leading up to an engineering crack with a length up to a few average grain diameters (roughly 100–200 \( \mu \)m for 304LSS). The crack initiation microstructural feature for each micro-crack was determined using EBSD. The nucleation of a micro-crack at a microstructural feature was determined for twin boundaries, grain boundaries, and MSC growth information for the stress amplitude of 330 MPa. A few average grain diameters (roughly 100–200 \( \mu \)m for 304LSS) was used to capture crack nucleation and MSC growth information for the stress amplitude of 330 MPa. A few average grain diameters (roughly 100–200 \( \mu \)m for 304LSS) was used to capture crack nucleation and MSC growth information for the stress amplitude of 330 MPa. A few average grain diameters (roughly 100–200 \( \mu \)m for 304LSS) was used to capture crack nucleation and MSC growth information for the stress amplitude of 330 MPa.
given orientation and the arrows highlight the direction of the crack initiation out of the void. Notice that each arrow is parallel to one of the (111) plane traces indicating the crack is likely initiating along a primary (111) < 110 > slip system in the FCC material.

4.1.2. Slip Bands (SB)

Very ductile metals with low stacking fault energy (SFE) can behave in a similar way to pure metals that do not have major inclusions in which cracks initiate primarily by slip. In these cases, cyclic
deformation occurs due to slip in individual grains leading to an intrusion/extrusion topography at the surface. It has been shown that very narrow closed crack nuclei typically develop at the bottom of these intrusions [32]. Austenitic stainless steels generally fall into this category of low SFE metals and as such the (111) < 110 > slip systems play an important role in the deformation behavior. The slip behavior was observed to increase during fatigue loading of the interrupted SGF specimens in which slip traces on the surface increased in density with many eventually resulting in slip traces in two directions. These slip bands are shown to be aligned with the (111) plane as indicated by the dark lines across the grains in Figs. 5 and 6. Nearly all grains containing slip traces on the surface had one primary slip system activated as indicated by the single direction of the slip band traces in Figs. 5 and 6. However, many grains did show two active slip systems as can be readily observed by the crossing slip traces in the upper left grain in Fig. 6.

Slip bands accounted for nearly 30% of crack initiations during the interrupted load control fatigue testing. Cracks that initiated transgranularly rather than intergranularly were almost exclusively along a (111) plane trace that contained a slip system with relatively large Schmid factors (> 0.45). These cracks were considered to be initiated by intrusion/extrusion as described by Neumann and Tonnessen [32]. In these cases, where slip bands were considered as the initiation mechanism, the cracks were mostly oriented on the surface at a large angle with the loading direction with more than half of them having an angle greater than 70° with respect to the loading direction. It should be noted, however, that observation of surface cracks gives only a two-dimensional representation. In addition, given the relatively high Schmid factors (> 0.45), the SB cracks are most likely not perpendicular to the surface.

Fig. 6. Inverse pole figure maps collected by EBSD showing crack initiation from an etched out δ-ferrite inclusion.

Fig. 7. EBSD maps for a grain boundary crack showing a) inverse pole figure, b) phase map, c) Schmid factor map, and d) Taylor factor map.
A technique described elsewhere [39] has been shown to give reasonable results when determining if a twin boundary is coherent or incoherent past the surface interpretation provided by EBSD analysis. In this case, the twin trace normal is plotted on a (111) pole figure for both grains adjacent to the twin boundary. If the twin trace normal coincides with the incident planes, then the twin can be called coherent with reasonable certainty. The Brandon criteria allows for slight deviations in misorientation such that twin trace normals that fall slightly off of the coincident planes can still be considered coherent. This method is shown in Fig. 8 where the (111) incident planes for the adjacent grains along the twin boundary are plotted on a (111) pole figure (loading direction is horizontal in respect to Fig. 8). The boundary trace normal is plotted from the center of the pole figure and falls within the vicinity of the coincident planes indicating the twin boundary is coherent. From this method, it was found that every crack initiated from a twin fell within the range given by the Brandon criterion to be considered coherent.

4.1.5. Crack nucleation summary

As discussed in the previous sections, crack nucleation occurred at many microstructural features. Fig. 9 details the frequency of each microstructural feature associated with the nucleation of the analyzed micro-cracks along with comparison of SFM, TFM, and crack initiation angles with respect to loading direction. Fig. 9(a) clearly indicates that, similar to the initiation mechanism for the high cycle fatigue regime can be an indication of dislocation impingement across the boundary, resulting in stress concentrations along the boundary [20,38]. While Fig. 7(c) shows a relatively small SFM of 0.04, it was shown by Zhang et al. [20] that the SFM effect was highly dependent on the SFE of the material. For an austenitic stainless steel with a SFE similar to that of 304LSS (15 mJ/m²), the range of SFM for the transition from SB to intergranular cracking should be within 0.02–0.05. As evident in Fig. 7(c), the SFM falls within the required range reported by Zhang et al. [17]. Interestingly, every grain along a cracked grain boundary that had a relatively low Schmid factor (< 0.4) had a high Taylor factor. This is interesting because research has also suggested that high mismatches between Taylor factors result in localized strain incompatibilities, which may cause intergranular cracking [21,38]. Fig. 7(d) shows a low-medium (LM) Taylor factor mismatch (TFM) between the neighboring grains that nucleated the crack. While not all grain boundary cracks had large TFM values, nearly every grain boundary crack had at least one high Taylor factor associated with it indicating the importance of the misorientation between grains in crack nucleation.

4.1.4. Twin Boundaries (TB)

Crack initiation at twin boundaries has been previously observed [2,3,21,39], and is somewhat surprising due to the low energies associated with twin boundaries. The elastic incompatibility on the surface of a twin pair results in high shear stress concentrations at the interface of these boundaries which can lead to crack initiation. For FCC materials, the 23 twins run parallel to a (111) slip plane for both grains along the twin boundary, and thus, local plasticity arises along a dominant slip system (i.e. (111) < 110 >) parallel to the twin [2]. Neumann and Tonnessen [32] showed that TBs cracked due to persistent slip bands near the high localized shear stresses at the TBs which had a modified Schmid factor higher than 0.44.

Twin boundaries showed the highest frequency in crack initiation for this 304LSS alloy where they accounted for nearly half of the crack initiation sites. Cracks that initiated at twin boundaries showed the tendency to occur along the plane of maximum shear stress (∼45° from loading axis), but quickly propagated into the adjacent grains along the plane of maximum principal stress. This result is in agreement with previous research in which simulations showed that TBs oriented at 45° from the loading direction were susceptible to early intergranular fatigue crack formation as a result of strain localization along the TB [40].

4.1.3. Grain Boundaries (GB)

As previously mentioned, GB are defined as any boundary with a misorientation to its neighbor grain larger than 5° and not considered a twin boundary. The vast body of available research on the influence of microstructure on fatigue behavior for many alloys indicates the importance of gaining a better understanding of the role of GB on the crack initiation [33–37]. For the 304LSS alloy investigated, grain boundary cracks were largely uncommon throughout the MSC specimens tested accounting for roughly 13% of all initiation features. A grain boundary crack is shown in Fig. 7 including (a) grain orientations, (b) phases, (c) Schmid factors, and (d) Taylor factors (loading direction is horizontal in respect to Fig. 7). In this image, there is clear evidence of more than one active slip system as indicated by the slip traces in the left most grain, one parallel and one at an angle to the grain boundary. Nearly all cracked grain boundaries were similar in that the misorientation between the two grains was large with many of the GB cracks ranging from 40–60°, agreeing with a previous report on the crack nucleation behavior [36]. GB with high misorientations may result in elastic incompatibilities that can lead to intergranular crack initiation [37]. A (111) plane associated with each grain was observed to be somewhat parallel with the boundary that initiated the crack as shown by the star bursts highlighting the (111) plane traces in Fig. 7(a). The fact that the grain boundary is nearly parallel with the slip plane trace can result in higher stresses due to dislocation pileup at the grain boundary. Interestingly, these grain boundary cracks were near twin misorientation falling just outside of the Brandon criterion. It is believed that these boundaries are so near the twin misorientation that they show similar characteristics to the twin boundary cracks, which are discussed later.

Previous research has shown that a large SFM of grain boundaries
TB are the dominant crack initiation feature for the transitional fatigue regime. However, SBs also show a significant initiation frequency compared to twin boundaries.

As previously mentioned, GBs that contain relatively large SFM values can be indicative of plastic incompatibility, in addition to any elastic incompatibilities. The added stress concentrations along the boundary can ultimately lead to intergranular crack initiation along the GB. Fig. 9(b) presents the SFM for grain and twin boundaries. Cracked twin boundaries showed on average lower SFM values with roughly 60% of them having a SFM of 0.02–0.04 as compared to grain boundaries with almost 67% of them having a SFM of 0.03–0.04. While the SFM for grain boundaries may appear to be relatively low as well, for materials with low SFE (such as 304LSS), it was shown that a lower difference in Schmid factors (0.02–0.04) would still promote intergranular cracking rather than slip band cracking. Additionally, high TFM are indicative of fatigue crack susceptibility due to strain incompatibilities along the boundary. Fig. 9(c) shows similar behavior when comparing TFM of twin boundaries and grain boundaries. Twin boundaries tended to initiate when a TFM was present, which is in agreement with previous research. The fact that GB cracks showed similar TFM across their boundary as TB cracks could be the result of many cracked GBs (approximately 40%) having misorientations very close to the Σ3 twin boundary misorientation, falling just outside of the Brandon criterion. Additionally, more than 50% of the GB cracks that had similar misorientations as TB showed similar crack initiation angles relative to the loading direction as TB cracks. It should be noted, however, that the sample size of GB cracks compared to TB and SB cracks was quite small.

Though TB and SB were observed to be the most predominant crack initiation feature for 304LSS in this study, the crack initiation mode between these features differs. Twin boundaries generally cracked as a result of strain localization which are triggered by strain incompatibilities along the boundaries with most of them being oriented on the surface around the direction of maximum shear (45°) as indicated in Fig. 9(d). Slip bands, however, initiated almost exclusively near perpendicular on the surface; however, again it must be noted that EBSD gives only a two-dimensional representation and the crack may not be fully perpendicular below the surface. In this case, the crack initiation behavior appears to be similar to the one described by Neumann and Tonnessen.

Grain boundaries showed a mixture of both 45° and 90° orientation from the direction of loading. The observed differences in the crack initiation angles for the most frequent initiation features highlights the significance of twin and grain boundary orientation on the fatigue crack initiation of 304LSS. These findings indicate that the fatigue behavior of 304LSS could be potentially improved by minimizing the number of twin boundaries within the material and by controlling their relative angles with respect to loading direction through proper manufacturing and post-manufacturing processes.

4.2. Microstructurally short crack growth behavior

Martensite was rarely able to be unambiguously determined as the source of crack initiation and as such showed the lowest frequency of all crack initiation features. In general, most initiation sites had very little transformation occurring until a crack was more than a grain diameter in length. Roth et al. also reported similar results for the high cycle fatigue behavior of a 304LSS alloy. All cracks showed development of α’ transformed areas near the crack tip. Typically, the α’ phase began to develop at the crack tip as the crack propagated to a length above 15–20 µm. As the crack grew through the microstructure and developed a larger plastic zone ahead of the crack tip, martensitic zone sizes grew as well. As for crack nucleation, martensite formation did not appear to play a significant role. However, as discussed in the introduction, crack initiation includes the life to nucleate a crack and the subsequent MSC growth up until a crack reaches a length of a few average grain sizes. Previous research has shown that fatigue crack growth rates are higher.
in the martensitic phase as compared to the austenitic phase [10]. For this reason, it was of particular importance to study the effect of the transformed α’ zone at the crack tip on the MSC rates.

An α’ zone size was calculated as the average diameter measured across the transformed portion of a crack tip as shown in Fig. 10 (the loading direction is horizontal with respect to the figure). Multiple measurements were taken perpendicular to the change in crack length, Δ2a, from the previous cycle. In many instances, the crack propagated into an area of previously transformed martensite which resulted in a large increase in zone size for the given measurement. A representation of one of these areas can be seen directly below the zone size measurement in Fig. 10. These areas were included in the average α’ zone measurements to account for their effects on the crack growth behavior. The measurements were then averaged to obtain an average zone size for the propagating crack at each interruption schedule along with the calculated standard error.

The crack growth data compared to α’ zone sizes is given in Fig. 11 where each data point is taken at the end of every interruption schedule for a single crack. For analysis, at least two and up to four cracks were included for each initiation type in order to obtain MSC growth behavior for multiple cracks as shown in Fig. 11. Additionally, all analyzed micro-cracks were far enough from other cracks as to not be influenced by their presence. The data set for Fig. 11 contains the largest crack observed for each initiation feature in order to compare the MSC growth rates for the fastest growing crack in each category. The α’ zone size is plotted versus crack length in Fig. 11(a) indicating an increase in α’ zone size with increasing crack length regardless of the initiation mechanism involved. This zone size increase can be attributed to the growing plastic zone at the crack tip promoting phase transformation around the crack front as described by Roth et al. [4]. It can be seen in Fig. 11(a) that the smallest α’ zone sizes for GB and TB cracks were similar in size to those for the largest slip band and inclusion initiated cracks. This is a result of GB and TB cracks having longer crack lengths (2a) for a given interruption schedule, and thus, maintaining larger α’ zone sizes throughout their associated fatigue life.

The size of the transformed zone around the crack tip showed some level of correlation to MSC growth rates up to crack lengths of 500 µm, as indicated in Fig. 11(b). Comparing Fig. 11(a) and (b), it can be observed that the increase in MSC growth rates coincide with increasing crack length. It can also be noticed from Fig. 11(b) that the α’ zone size for GB and TB cracks tended to develop at a much higher rates compared to slip band and inclusion initiated cracks. Again, this is a result of GB and TB cracks having longer lengths as compared to other microstructural features for a given cycle count, indicating that they are susceptible to earlier crack initiation compared to other microstructural features. What is interesting, however, is the similarity of many GB to TB cracks including their orientation, nucleation behavior, and MSC growth behavior. It was previously mentioned that many GB cracks showed a similar misorientation across the boundary to cracked TBs, with just enough deviation to fall outside of the Brandon criteria for Σ3 twin boundaries. The GB cracks that were similar to TB cracks showed approximately 33% longer crack lengths than other GB cracks for a given cycle. These GB cracks had very similar crack lengths and growth rates as TB cracks, as shown in Fig. 11. Additionally, GB and TB cracks were 50–65% longer than other cracks for a given cycle, further indicating that these types of cracks initiate earlier and as a result grow faster than other microstructural features. These results show that crack initiation life for a transition fatigue regime is more sensitive to crack initiation feature than the martensitic transformation zone that occurs at the crack front. Specifically TB cracks and GB cracks with similar misorientations to twin boundaries were observed to have much longer lengths than other microstructural features for a similar number of cycles indicating that these cracks nucleated and developed faster than other microstructural features. As such, grain and twin boundary cracks reach the end of the crack initiation stage well before cracks from other initiation mechanisms, and most likely, are the initiation features to form a dominant crack to failure.

5. Conclusions

Through the use of SEM and EBSD, several micro-cracks were observed and analyzed to understand the initiation mechanisms in the transitional fatigue life regime. The effect of γ → α’ phase transformation on the crack initiation behavior was also investigated. The following conclusions can be made regarding the nucleation and MSC growth of a micro-crack of 304L stainless steel:

1. Martensitic transformation was found to occur during fatigue loading at 330 MPa, which is within the transition fatigue life regime. Localized plasticity at the crack tip results in martensite...
formation around the crack tip for cracks reaching 10–50 µm.  

2. For fully reversed fatigue loading within the transition fatigue life regime, twin boundaries were the leading crack initiation feature. However, other microstructural features, such as slip bands and grain boundaries, contributed to crack initiation as well.

3. Slip bands were shown to be the second most frequent crack initiation feature. Slip band cracks, however, reached the end of the crack initiation stage later in the fatigue life as compared to twin and grain boundary cracks. Cracks that initiated from twin and grain boundaries were roughly 66% longer than cracks that initiated from slip bands. Therefore, twin and grain boundary cracks will most likely be the ones to grow to failure.

4. Twin boundaries predominately cracked at an angle close to 45° to the loading direction while slip bands were nearly perpendicular (90°) to the loading direction. Grain boundary cracks showed a tendency to initiate at both 45° and 90° in relation to the loading direction.

5. Martensite zone size at the crack tip was found to correlate to the crack length in the crack initiation stage as a result of the developing plastic zone as the crack grows in size. Grain and twin boundary cracks were found to initiate at an earlier stage of fatigue life and develop larger α' zone sizes in the initiation stage.

6. Crack initiation life was found to be much more sensitive to initiation feature than martensite formation near the crack. Specifically, twin boundaries showed the greatest frequency of all crack initiation features and were the most likely to lead to a dominant crack. Additionally, grain boundaries with similar misorientations as twin boundaries initiated cracks in a similar manner to twin boundary cracks.

In summary, the crack initiation behavior of 304L stainless steel is dominated by twin boundaries in the transitional fatigue regime. While other microstructural features such as slip bands showed significant initiation sites, the MSC growth from these features was slow as compared to that of TB and GB micro-cracks. Although limited in number, GB micro-cracks that initiated in a similar fashion as TB and grew at similar rates. Finally, these results show that the crack initiation life could be improved by limiting the number of TB or controlling the direction of them in relation to the loading direction of the material through manufacturing and/or post manufacturing processes.

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References