The Mechanism of Low-cycle Fatigue Crack Formation Related to Annealing Twin Boundaries in Austenitic Stainless Steels*

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Synopsis

Crack along annealing twin boundary lines can frequently be observed on the surface of low-cycle-fatigued specimens of austenitic stainless steels. In this study, a mechanism for the formation of such fatigue cracks is proposed.

When cracks are initiated along twin boundary lines in a crystal grain which contains many twin boundary planes parallel to each other, the cracks are initiated only along alternate twin boundary lines. These cracks do not propagate along the twin boundary planes, but deviate from the twin boundary plane with increasing depth.

The mechanism of the formation of these cracks is discussed from a crystallographic point of view. Such cracks are initiated in crystal grains where the strain concentrated on a $\{111\}$ plane including a twin boundary line at the surface. Two slip systems along $\{111\}$ planes, one of which is parallel to the twin boundary plane, are then activated alternately. The crack propagates approximately along a $\{101\}$ plane, which is an intermediate plane between the $\{111\}$ planes.

I. Introduction

The process of fatigue fracture in metals can generally be divided into two parts; initiation of the fatigue cracks and their subsequent propagation. The mechanism of crack initiation depends on the structure of the material, the type of loading, temperature, environment, etc. For high-cycle fatigue cracks in ductile materials, it has been frequently observed that the cracks initiate along planes on which slip has occurred.¹⁾ For materials containing annealing twins, high-cycle fatigue cracks have been observed along twin boundaries,²⁾ which are also considered to be formed by slip processes along the twin boundary planes.³⁾

The authors have carried out systematic high-temperature, low-cycle fatigue tests on austenitic stainless steels.⁴⁾ Observations of the surfaces of fatigued specimens have shown that some fatigue cracks initiate along annealing twin boundary lines.⁵⁾ In this study, the mechanism of the formation of such low-cycle fatigue cracks is discussed from a crystallographic point of view.

II. Review of Previously Published Results

A summary is first given on the previously published experimental results for the mechanisms of the formation of fatigue cracks in ductile materials, and the role of annealing twin boundaries in the formation of fatigue cracks.

Under cyclic stress, slip bands nucleate along slip planes in materials without any sharp stress concentrators being required, and intrusions and ex-

trusions then form along the slip bands.⁶⁾ These intrusions expand and become cracks along the slip bands. Forsyth⁷⁾ called this stage of initiation and crack propagation Stage I, and distinguished it from Stage II, in which the crack propagates in a plane perpendicular to the direction of maximum tensile stress.

As an alternative to the model of fatigue crack formation along planes of one slip system, Neumann⁸⁾ proposed a model for the formation of a fatigue crack by the coarse slip of two slip systems, in which the macroscopic plane of the crack does not coincide with the slip planes.

It has been recognized that twin boundaries are favoured sites for the formation of fatigue cracks in face-centered cubic (FCC) metals.²⁾ Boettner *et al.*⁹⁾ studied fatigue crack formation at twin boundaries in copper as a function of the twin boundary orientation with respect to the applied stress, and obtained the following results:

(1) When the primary slip plane is parallel to the twinning plane and the slip vector has an angle to the specimen surface, a prerequisite for fatigue crack formation at twin boundaries is that slip occurs on planes parallel to the twin boundary.

(2) When the primary slip plane is parallel to the twinning plane and the primary slip direction lies along the specimen surface, cracks are formed at only alternate twin boundaries, and the operation of secondary slip planes facilitates crack formation.

(3) Fatigue crack formation at twin boundaries requires the operation of slip near the twin boundaries.

Fatigue cracks initiated along annealing twin boundary lines were observed by the authors at the surfaces of low-cycle-fatigued specimens of austenitic stainless steels.⁵⁾ Photograph 1 shows an example of such cracks. As these cracks appeared not to lie on the twin boundary planes,⁵⁾ the mechanism for cracks formation along twin boundary planes by slip processes is not applicable.

III. Materials and Experimental Procedure

The materials studied were solution treated Types 304 (18Cr-9Ni), 316 (17Cr-10Ni-2Mo) and 321 (17Cr-9Ni-0.3Ti) stainless steels. In these materials, annealing twins are frequently observed. In the following description, the term "twin boundaries" means these annealing twin boundaries.

The gage lengths of the specimens were polished

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with 600 grade silicon carbide paper in the longitudinal direction and fatigue tested on tension-compression, servo-controlled, hydraulically actuated fatigue testing machines. During testing, specimens were heated by an electric resistance furnace which completely surrounded them.

In these materials, fatigue cracks along twin boundary lines were observed on the surface of specimens tested at temperatures below 600°C.⁵⁾ The surfaces and cross-sections of specimens were observed in detail in order to clarify the relation between the cracks and the twin boundary planes. After the surfaces of specimens were polished with 600 grade silicon carbide paper to remove the oxide film, saturated oxalic acid solution electro-etching was used to reveal structures and cracks. Etch pitting of specimen surfaces was used to determine the crystallographic orientations, the etching reagent being 5 to 30 cc HCl, 30 cc H₂O₂ and a little HF. Etch pits and fracture surfaces were observed in a scanning electron microscope.

After etching, fine lines corresponding to slip bands formed during mechanical cutting could be observed, as shown in Photo. 1, in grains on the surfaces of the specimens. These lines were found in untested specimens also. As it was difficult to determine unambiguously the relation between the crystallographic orientation of a crystal in which a crack was observed along twin boundary and the loading direction, two parameters θ and α were measured at surfaces of specimens as shown in Fig. 1.

Photo. 1. Optical micrograph of a specimen surface showing fatigue cracks along twin boundary lines. 316 steel fatigued at 450°C at a strain rate of $6.7 \times 10^{-4} \text{ s}^{-1}$ and a total strain range of 2×10^{-2} .

Fig. 1. Definition of θ and α , the angles made by the twin boundary line and the direction perpendicular to the loading axis, and slip bands, respectively.

These angles made by the twin boundary line and the direction perpendicular to the loading direction, and by the twin boundary line and slip bands due to mechanical cutting, respectively.

IV. Results of Observation

1. Cracks Observed on Surfaces of Specimens

As shown in Photo. 1, when cracks initiated along twin boundaries in a crystal grain which contained many twin boundary planes parallel each other, cracks were found only along alternate twin boundary lines. Photograph 2 shows another example of such cracks. For a crystal grain containing twin boundary planes which are not parallel, cracks were found along neighbouring twin boundary lines, as shown in Photo. 2 (a). Many cracks, as shown in Photos. 1 and 2 (a), followed twin boundary lines continuously. Several discontinuous cracks along one twin boundary line could also be found, as shown in Photo. 2 (b). The ends of each crack deviated from the twin boundary line, after which they seemed to connect up with each other.

The angles θ and α defined in Fig. 1 were measured in observations of many cracks along twin boundary lines. The results are shown in Figs. 2 and 3. The

(a) The same specimen as that shown in Photo. 1
(b) 321 steel fatigued at room temperature at a strain rate of $6.7 \times 10^{-4} \text{ s}^{-1}$ and a total strain range of 2×10^{-2}

Photo. 2. Optical micrographs of a specimen surface showing fatigue cracks along twin boundary lines.

Fig. 2. Distribution of the angle θ made by twin boundary lines along which a fatigue crack formed and the direction perpendicular to the loading axis.

Fig. 3. Distribution of the angle α made by the twin boundary lines along which a fatigue crack formed and slip bands due to mechanical cutting.

angle θ is distributed symmetrically with respect to 0° within the angles 45° to -45° . The distribution has a peak in the range 30° to 40° . The angle of α is distributed from 25° to 75° , being concentrated from 55° to 75° .

2. Cracks Observed on Specimen Cross Sections

No crack could be observed along twin boundary lines on cross sections. In Photo. 3, cracks that initiated along twin boundary lines at the surface of the specimen did not propagate along the twin boundary plane but deviated from them with increasing depth. Etch pits observed in region A of the surface shown in Photo. 3 (a) are shown in Photo. 3 (b). A schematic explanation of Photo. 3 is given in Fig. 4, showing angles a , b , c , d , and e , and indicating their values.

Discontinuous cracks, as shown in Photo. 2 (b), did not propagate into the interior of the specimen. However, at the stage where the discontinuous cracks connected up with each other, they were able to propagate into the interior, deviating from the twin boundary plane. The result that some of the cracks shown in Photos. 1 and 2 do not totally lie on twin boundary lines is considered to be due to the observations being carried out after polishing the surface of the specimens to remove the oxide film.

3. SEM Observations of the Fracture Surfaces

Since the cracks discussed above were not the main cracks which led to complete failure of the specimen, it was difficult to observe their fracture surfaces. Photograph 4 shows an example of a crack along a twin boundary line at the surface which had propagated until complete fracture of the specimen occurred. Fatigue testing of this specimen was carried out after showing the structure by etching. In the

(a) Optical micrographs of the surface and the cross section first observed
(b) SEM micrograph of region A of the surface, showing etch pits

Photo. 3. Observation of the surface and cross section of a specimen of 304 steel fatigued at 600°C at a strain rate of $6.7 \times 10^{-3} \text{ s}^{-1}$ and a total strain range of 2×10^{-2} .

Fig. 4. Schematic illustration of the surface and cross section of the specimen shown in Photo. 3.

- (a) Optical micrograph of the surface of the specimen
 (b) SEM micrograph of the fracture surface for the initial stages of crack formation

Photo. 4. Observations of the surface and the fracture surface of a specimen of 316 steel fatigued at room temperature at a strain rate of $6.7 \times 10^{-3} \text{ s}^{-1}$ and a total strain range of 1×10^{-2} .

photograph of the surface of the specimen (Photo. 4 (a)), a small crack seems to have been initiated on the alternate twin boundary line. On the fracture surface, some regions of the initial crack that propagated from the specimen surface were smoother than the neighbouring regions. Under higher magnification, fine straight striations which had a common direction were observed in these smooth regions (Photo. 4 (b)).

V. Discussion

1. Model for Fatigue Crack Formation Related to Annealing Twin Boundaries

The arrangement of atoms around a twin boundary in FCC metals is symmetrical with respect to a $\{111\}$ plane, the coherent twin boundary. This was confirmed in the present experiment in a thin foil specimen by electron diffraction analysis. This twinning relation is shown in Fig. 5 as a combination of two regular octahedrons whose facets are $\{111\}$ planes. In the following discussion, the crystal on one side of the twin boundary is called the parent

Fig. 5. Twinning relation in FCC metals, shown by the combination of two regular octahedrons whose facets are $\{111\}$ planes.

matrix M , and its crystallographic orientations are indicated by Miller indices with the suffix M . The crystal on the other side is called the twin matrix T , and its crystallographic orientations are indicated by Miller indices with the suffix T . Each orientation in this twinning relation is indicated in Fig. 5, where $(111)_M$ and $(111)_T$ planes are in contact with each other at the twin boundary plane.

From observations of the directions of twin boundary lines at the surface and on the cross section of the specimens, together with observations of etch pits and the relation between the loading direction and slip bands due to mechanical cutting, the orientation of the crystal grain shown in Photo. 3 can be determined approximately to be as shown in Fig. 6, where the loading direction coincides with the $\langle 110 \rangle_T$ direction and the free surface is the $(111)_T$ plane. From the relation between the crack plane and the twin boundary plane, it is considered that the crack plane coincides with the $(101)_M$ plane approximately. The angles shown in Fig. 4 are in good correspondence with those shown in the crystal model in Fig. 6.

In such a crystal model there are three $\{111\}$ planes, each of which includes the twin boundary line at the surface of specimen, these being $(111)_T$, $(111)_M (\equiv (111)_T)$ which is the twin boundary plane itself, and $(111)_M$. As the plane $(111)_T$ is parallel to the loading direction, slip can not occur on this plane. Slip along $(111)_M \langle 011 \rangle_M (\equiv (111)_T \langle 101 \rangle_T)$ has a Schmid factor of $1/\sqrt{6}$, and is considered to occur easily. Boettner *et al.* pointed out⁹⁾ that non-coherent steps along twin boundaries can act as dislocation sources. However, if only slip along the twin boundary plane is an important factor in the initiation of fatigue cracks, it is difficult to explain the result that cracks initiate regularly along alternate twin boundary lines. Slip along $(111)_M \langle 011 \rangle_M$ has a Schmid factor of $10/9\sqrt{6}$ and is also considered to occur easily. In region A shown in Fig. 6, slip can take place to the free surface, but in region B slip is interrupted by the twin boundary plane. In spite of having one slip system, slip does not occur uniformly in the parent matrix M . Because of this irregularity, it is considered that strain is concentrated at the boundary of the A and B regions, which is the $(111)_M$ plane, including the twin boundary line at the surface. This slip system is considered to play

an important role in the initiation of fatigue cracks at twin boundary lines at the surface.

If the mechanism of fatigue crack formation along a slip plane is applicable to this situation, the fatigue crack must lie on the $(1\bar{1}1)_M$ plane. This contradicts the observations that the crack plane coincides with the $(101)_M$ plane approximately. From this discussion, it can be seen that the fatigue cracks shown in Photo. 3 were not formed by the action of one slip plane alone. However, the model for the coarse slip of two slip systems proposed by Neumann⁸⁾ may be applicable to the formation of fatigue cracks related to annealing twin boundaries.

As shown in Fig. 7, slip can occur easily not only along $(1\bar{1}1)_M<0\bar{1}\bar{1}>_M$ but also along $(1\bar{1}\bar{1})_M<0\bar{1}1>_M$, which is parallel to the twin boundary plane, in the parent matrix M . If these two slip systems are activated alternately, a crack can propagate along an intermediate plane $(101)_M$. This coincides with the results of observation. As shown in Fig. 7 (b), in practice slip occurs along many $(1\bar{1}1)_M$ and $(1\bar{1}\bar{1})_M$ planes at the crack tip under tensile loading, and this slip is considered not to be exactly reversed during compressive loading, due to oxidation at the crack tip, etc. This process is the same as that for the formation of fatigue striations in FCC metals proposed by Pelloux.¹⁰⁾ The surface of a crack formed by the

model shown in Fig. 7 must therefore have striations. From low-cycle fatigue tests under high cyclic stress, it is considered that fatigue cracks can be formed by the activation of two slip systems.

2. Relation between the Direction of the Loading Axis and the Schmid Factor

The above discussion is based on the crystallographic configuration shown in Fig. 6. However, the actual relations between the loading direction and orientations of crystals do not always coincide with this configuration. The way in which Schmid factors for slip relating to the formation of fatigue cracks change with respect to the relation between the direction of the loading axis and the orientations of crystals will therefore be discussed in detail.

Taking $\langle 21\bar{1} \rangle_T$ as a basic direction, the relation between this direction and the loading direction can be expressed by θ_1 and θ_2 , as shown in Fig. 8, which are angles between the loading direction and the $(1\bar{1}1)_T$, and $(011)_T$ planes, respectively. The angles θ_1 and θ_2 for the loading direction in Fig. 6 are 0° and 30° , respectively. The Schmid factors for slip on two pairs of slip systems can be calculated, i.e., for the slip systems $(1\bar{1}1)_M<0\bar{1}\bar{1}>_M$, whose Schmid factor

- (a) The combination of two regular octahedrons seen from the surface of $(1\bar{1}1)_T$ plane
- (b) The cross-section AA, perpendicular to the $(1\bar{1}1)_T$ plane
- (c) The shapes of etch pits on the surface

Fig. 6. Crystallographic configuration for the crystal grain shown in Photo. 3.

Fig. 7. Mechanism of the formation of fatigue cracks along twin boundary lines at the specimen surface.

Fig. 8. Definition of θ_1 and θ_2 , expressing the relation between the loading direction and the crystallographic orientation of a crystal grain with a twin boundary plane.

is designated by AI, and $(\bar{1}\bar{1}\bar{1})_M \langle 0\bar{1}1 \rangle_M$, designated by BI, and for the slip systems $(1\bar{1}1)_M \langle \bar{1}\bar{1}0 \rangle_M$, designated by AII, and $(\bar{1}\bar{1}\bar{1})_M \langle 1\bar{1}0 \rangle_M$, designated by BII. It is considered that the mechanism for the formation of a crack shown in Fig. 7 is applicable when the Schmid factors for either of the two pairs of slip systems are large enough for slip to occur.

Figure 9 shows the results for $\theta_1 = 7^\circ 20'$ ($k=0.75$) and $\theta_1 = 15^\circ 47'$ ($k=0.5$), the parameter k being defined together with its calculation method described in the Appendix. The curves for AI and BI in the negative region of θ_2 have the same shape as the curves for AII and BII in the positive region, and *vice versa*. Contour lines within which both AI and BI values are larger than 0.35, 0.4 and 0.45 are shown in the Fig. 10 θ_1 - θ_2 diagram. The contours for AII and BII are located symmetrically with respect to those for AI and BI about zero θ_2 . The point corresponding to a θ_1 of 0° and a θ_2 of 30° in Fig. 10 shows the results for the situation shown in Fig. 6. Therefore, when

the loading direction inclines from the position in Fig. 6 by 20° to 30° , the Schmid factors for two slip systems which can contribute to form a fatigue crack by the mechanism shown in Fig. 7, can have larger values than 0.4.

3. Discussion on the Experimental Results

1. Distributions of θ and α Shown in Figs. 2 and 3

If it can be shown that fatigue cracks can be formed easily in crystals having large Schmid factors for the two slip systems, for example, larger than 0.4, it would be confirmed that the model for the formation of fatigue cracks shown in Fig. 7 is an appropriate one. The values of θ_1 and θ_2 could not be determined experimentally. Therefore in this section, the values that θ and α , which can be measured, should have is discussed for the case where fatigue cracks are formed in crystals having large Schmid factors for the slip systems shown in Fig. 7, and the results are compared with the experimental results shown in Figs. 2 and 3, where the values of α were measured for the slip bands observed in the twin matrices.

When the values of θ_1 and θ_2 change while maintaining the $\langle 011 \rangle_T$ ($\equiv \langle \bar{1}01 \rangle_M$) direction on the twin boundary plane parallel to the free surface, the angle α varies with θ_1 as shown in Fig. 11. When the crystal rotates around the loading direction, the Schmid factors do not change but the values of θ and α vary. For example, for a θ_1 of 0° and a θ_2 of 30° , both α and θ change their values with the rotation around the loading direction. The results of this are shown in Fig. 12. In this case, however, the $\langle 011 \rangle_T$ ($\equiv \langle \bar{1}01 \rangle_M$) direction does not lie on the

Fig. 9. Change in the Schmid factors for slip systems that contribute to crack formation.

Fig. 11. Change in α with respect to the change in θ_1 .

Fig. 10. Contour lines for both Schmid factors AI and BI larger than 0.35, 0.4, and 0.45.

Fig. 12. Changes in θ and α due to rotation of the configuration of the crystals shown in Fig. 10, i.e., in the case of $\theta_1 = 0^\circ$ and $\theta_2 = 30^\circ$.

free surface, and discontinuous cracks can be initiated along the twin boundary line at the surface. This phenomenon will be discussed later.

When the $\langle 011 \rangle_T$ direction is parallel to the free surface, where θ has the same value as θ_2 , the Schmid factor is large for values of θ_2 ranging from -45° to $+45^\circ$. In particular, for θ_2 around -30° and $+30^\circ$ the Schmid factor has a large value for a wide range of θ_1 . The distribution of θ is thus considered to have peaks around -30° and $+30^\circ$. The Schmid factor has a large value for θ_1 values ranging from 0° to 30° . If a fatigue crack is formed for that range of θ_1 , the value of α ranges from 55° to 60° , as shown in Fig. 11.

When the $\langle 011 \rangle_T$ direction does not lie on the free surface, the values of α and θ vary even for fixed values of θ_1 and θ_2 . For example, when the configuration shown in Fig. 6 rotates around the loading direction, the value of θ varies in the range 10° to 40° and the value of α can be smaller or larger value than 55° to 60° (Fig. 12).

In general, it is considered that orientations of crystal grains with twin boundaries are distributed at random with respect to the loading direction. From the above discussion, if fatigue cracks are formed in crystal grains in which the Schmid factors of two slip systems have a large value, for example larger than 0.4, it is considered that the distribution of the values of θ observed will be located in the range from -45° to $+45^\circ$, placed symmetrically about 0° , with peaks at about -30° and $+30^\circ$, and that the value of α will be distributed around 60° . These estimates coincide with the experimental results shown in Figs. 2 and 3.

2. Cracking Only at Alternate Twin Boundaries

When cracks were formed along twin boundaries in a crystal grain which contains many twin boundary planes parallel to each other, the cracks only initiated at alternate twin boundary lines, at the free surface. This phenomenon can be understood from the crystallographic configuration shown in Fig. 13. In this model, both parent and twin matrices repeat alternately, the relation between the matrices and the loading axis being the same as that shown in Fig. 6. The fatigue cracks are thus initiated only at alternate twin boundary lines, at the free surface.

3. Cracking in a Crystal Containing Non-parallel Twin Boundary Planes

One example of a crystal which contains non-parallel twin boundary planes is shown in Fig. 14(a), using combinations of regular octahedrons. The cross-section AA is also shown in Fig. 14(b). This model is symmetrical about a $(10\bar{1})_T$ plane. When the loading direction coincides with the $\langle 10\bar{1} \rangle_T$ direction, the left part of the model is identical with the crystallographic configuration shown in Fig. 6. Fatigue cracks can thus be initiated, from their symmetry with respect to the $(10\bar{1})_T$ plane, to which the loading direction is perpendicular, along the right part of the twin boundary line at the free surface, as well as the left part. Crack propagation to the interior is shown in Fig. 14(b).

Fig. 13. Mechanism of crack formation along alternate twin boundary lines only.

Fig. 14. Mechanism of crack formation in crystal grains containing non-parallel twin boundary planes

The cross section of the specimen shown in Photo. 2 (a) could not be observed in detail, but from observations of the slip bands due to mechanical cutting it is considered that the formation of the cracks shown in Photo. 2 (a) corresponds to the model. If the dotted line is the free surface in Fig. 14 (b), fatigue cracks must be initiated along twin boundary lines between which there are two twin boundary lines at the free surface.

4. Initiation of Discontinuous Cracks along a Twin Boundary Line at the Free Surface

The above discussion considers the situation in which one $(1\bar{1}1)_M$ plane comes into contact with the twin boundary plane along the twin boundary line at the free surface, as shown in Fig. 6 and Fig. 15 (a). If the grain is rotated with respect to the loading direction, a part of one $(1\bar{1}1)_M$ plane comes into contact with the twin boundary below the free surface, as shown by the dash-dotted line in Fig. 15 (b), and the other part of the $(1\bar{1}1)_M$ plane comes into contact with the free surface, as shown by the fine solid line. For the latter part of the $(1\bar{1}1)_M$ plane, slip takes place against the free surface, but for the former part slip is interrupted by the twin boundary plane. In spite of the one slip plane, slip does not occur uniformly. Because of this irregularity, strain concentrates near the boundary, which is shown by a solid line. At above or below the $(1\bar{1}1)_M$ plane, for example the $(1\bar{1}1)_M'$ or $(1\bar{1}1)_M''$ plane, the same phenomenon occurs as that at the $(1\bar{1}1)_M$ plane. By combination of such slip action and slip along the $(\bar{1}\bar{1}1)_M$ plane which is parallel to the twin boundary plane, cracks could be formed. Such cracks are not initiated

Fig. 15. Mechanism for the formation of discontinuous cracks at a twin boundary line.

along the twin boundary line uniformly, but must be initiated discontinuously as shown by the solid lines in Fig. 15 (b). The situation for these discontinuous cracks is an early stage of crack initiation, and such cracks seem not to propagate into the interior.

5. Fractography

The crystal orientation of the grain whose fracture surface is shown in Photo. 4 was not determined in detail, however, the crack seems to have been formed by the model shown in Fig. 7 since another crack was about to initiate along an alternate twin boundary line. It is expected that a crack formed using this model would have striations. In fact, the crack shown in Photo. 4 does have straight striations.

Pelloux¹⁰⁾ pointed out two crystallographic planes, i.e. {001} and {011}, which are intermediate planes between two {111} planes, as fracture planes on which striations can be formed. In his observations, however, fracture planes were only close to a {001} orientation. Neumann¹¹⁾ also pointed out that fracture surfaces can be formed in FCC metals along {001} planes, since the macroscopic crack front are usually bent. In the low-cycle fatigue of austenitic stainless steels, fracture surfaces of {001} orientation were observed, on which striations were formed.¹²⁾ Fracture surfaces formed by the model shown in Fig. 7 are close to a {011} plane. The fracture surface shown in Photo. 4 is thus considered to be close to a {011} plane, with striations.

VI. Conclusion

Cracks along annealing twin boundary lines were observed on the surface of low-cycle fatigued speci-

mens of austenitic stainless steels, and a mechanism for the formation of such fatigue cracks was proposed from a crystallographic point of view.

When cracks are initiated along twin boundary lines in a crystal grain containing many parallel twin boundary planes, cracks initiate only along alternate twin boundary lines. These cracks do not propagate along the twin boundary planes, and deviate from the twin boundary plane with increasing depth.

Such cracks are initiated in crystal grains where the strain concentrates at a {111} plane that includes the twin boundary line at the surface. Two slip systems along {111} planes, one of which is parallel to the twin boundary plane, are then activated alternately. The crack propagates approximately along a {101} plane, which is intermediate to the {111} planes.

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Appendix

When the loading direction is inclined by θ_1 , it lies on a $(\frac{1}{k}\frac{1}{k})_T$ plane, where

$$\cos \theta_1 = (2+k)/\sqrt{3(k^2+2)}$$

Any arbitrary direction on the $(\frac{1}{k}\frac{1}{k})_T$ plane can be described by $\langle 1 a a-k \rangle_T$. The direction for a θ_2 of 0° is $\langle 1 \frac{k}{2} \frac{k}{2} \rangle_T$. Therefore the relation between θ_2 and a on the $(\frac{1}{k}\frac{1}{k})_T$ plane is expressed as follows:

$$\cos \theta_2 = \sqrt{k^2+2}/\sqrt{2(k^2+2a^2-2ak+1)}$$

The loading axis $\langle 1 a a-k \rangle_T$ in the twin matrix is identical to the axis $\langle 2k-3a+2 \ 2k-1 \ -k+3a+2 \rangle_M$ in the parent matrix. The Schmid factors AI, AII, BI and BII can thus be written using parameters a and k as follows:

$$AI = |(k-5)(3a+k+1)/9\sqrt{6(k^2+2a^2-2ak+1)}|$$

$$AII = |(k-5)(3a-4k-1)/9\sqrt{6(k^2+2a^2-2ak+1)}|$$

$$BI = |(k+1)(a-k+1)/\sqrt{6(k^2+2a^2-2ak+1)}|$$

$$BII = |(k+1)(a-1)/\sqrt{6(k^2+2a^2-2ak+1)}|$$