Influence of small defects and nonmetallic inclusions on the high and very high cycle fatigue strength of an ultrahigh-strength steel

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Abstract
The high and very high cycle fatigue (VHCF) properties of ultrahigh-strength Ck45M steel processed by thermomechanical rolling integrated direct quenching were investigated. S–N tests with smooth and small drilled holes containing specimens as well as near-threshold fatigue crack growth measurements were performed up to $2 \times 10^{10}$ cycles using ultrasonic-fatigue testing technique. The fatigue strength of smooth specimens is mainly determined by the size of nonmetallic inclusions. For surface defects larger than 80 μm, the fatigue limit can be correlated with a constant threshold-stress intensity factor. The $\sqrt{\text{area}}$-parameter model adequately predicts the fatigue limit for internal defects and for surface defects with sizes between 30 and 80 μm. VHCF failures from smaller surface defects occur at stress amplitudes below the predicted fatigue limit. The long-crack threshold in ambient air is close to the effective threshold stress intensity factor. In optically dark areas at interior inclusions, cracks grow at mean propagation rates of $10^{-15}$ m/cycles.

KEYWORDS
defect sensitivity, fatigue limit, optically dark area, thermomechanically processed steel, threshold-stress intensity factor, ultrasonic fatigue

1 | INTRODUCTION

Very high cycle fatigue (VHCF) failure is an important issue for high-strength steels. Due to their low defect tolerance, fatigue cracks can even initiate at small inherent defects such as nonmetallic inclusions. Theoretical maximum fatigue strength—related with the matrix material in the absence of detrimental defects—can be only achieved in the case of super-clean materials and by avoiding component production-related flaws (e.g., insufficient surface finish, scratches, and punch marks). For steels, the upper boundary of the fatigue limit depends on the material hardness (e.g., the Vickers hardness $HV$) and can be estimated by $1.6 \cdot HV^\alpha$—which could be demonstrated even for high-strength steels with tensile strengths up to 1.7 GPa. In this case, however, the critical defect size is only a few microns, and larger flaws would already decrease the fatigue strength. While it may be possible to produce small defect-free fatigue test specimens, the probability of the presence of...
detrimental, fatigue-strength decreasing defects increases for large components or production series. This important issue should be considered when laboratory fatigue test results are used for component designs. Bearing in mind that fatigue failure potentially originates from the largest flaw, defect screening of the deployed material in combination with extreme-value rating is useful to predict the maximum defect in relevant, cyclically stressed volumes.\textsuperscript{1,5} Furthermore, fatigue tests with materials containing an intentionally increased number and size of defects may serve to trigger relevant fracture mechanisms, such as inclusion initiated fracture under cyclic torsional loading.\textsuperscript{6,7} To systematically characterize the defect sensitivity of a material, defects of different sizes should be utilized to determine the transition size between small and large defects. This is important from a fracture mechanics point of view since the threshold-stress intensity factor for small cracks and defects is size dependent, while it is a constant value for long cracks and large defects.\textsuperscript{4,8}

Several investigations on the VHCF properties of ultrahigh-strength steels—that is, steels with tensile strengths exceeding 2 GPa—have been performed (e.g., Refs.\textsuperscript{9–13}), the vast majority with bearing steels. These studies mainly focus on fatigue fracture emanating from interior inclusions, and several models have been proposed to explain failure at very high number of load cycles.\textsuperscript{14–19} A relatively seldom considered, but important, aspect with respect to VHCF is that typical low- and medium-alloy ultrahigh-strength steels exhibit rather poor corrosion resistance. This may lead to an elimination of the classical fatigue limit and can cause surface failures above $10^7$ cycles\textsuperscript{20} (in the present work, the fatigue limit is generally considered as the maximum stress level below which the lifetime is infinite). Close inspection of fatigue data obtained from tests with ultrahigh-strength steels indeed reveals that surface fracture even occurs in the VHCF regime.\textsuperscript{9,15,21,22} Consequently, VHCF failure of steels with tensile strengths in the range of 2 GPa and higher should not be considered as an issue exclusively associated with interior fracture.

In the present work, ultrasonic fatigue tests were performed with thermomechanically processed (TMP) and subsequently direct quenched (DQ) Ck45M steel (TMP-DQ). The chemical composition determined by spark optical emission spectroscopy is 0.440C-0.667Mn-0.294Si-0.009P-0.036S-0.218Cr-0.221Cu-0.0002B-0.034Mo (weight %). A $200 \times 80 \times 60$ mm block machined from the casting was soaked at $1150^\circ$C for 1 h and thermomechanically rolled in two stages, as shown in Figure 1. Stage 1 comprised hot rolling in three passes above the no-recrystallization temperature ($T_{nr}$) to a thickness of $\sim 31.5$ mm with $\sim 0.2$ strain/pass. Stage 2 comprised controlled rolling in the $T_{nr}$ regime (below $\sim 950^\circ$C) comprising three passes ($\sim 0.2$ strain/pass) to a thickness of $15.5$ mm with the finish rolling temperature at $\sim 830^\circ$C (above the ferrite start temperature, $A_r3$). Immediately after hot rolling, the sample was DQ in water to room temperature. Tensile tests were performed according to ASTM E8M standard.

Specimen shapes used for $S-N$ and near-threshold FCGR tests are shown in Figure 2A,B, respectively. All specimens were machined with their longitudinal axis in rolling direction. The surface in the gauge section of machined specimens was ground with emery paper up to

\begin{figure}[h]
\centering
\includegraphics[width=0.5\textwidth]{schematic_diagram.png}
\caption{Schematic diagram of the laboratory rolling schedule}
\end{figure}

was determined. Fracture mechanisms for both surface and interior failure are discussed considering environmental influences. In addition, the experimental results were evaluated applying fracture mechanics-based prediction models.

2 | MATERIAL AND EXPERIMENTAL PROCEDURE

Fatigue tests were performed with thermomechanically processed (TMP) and subsequently direct quenched (DQ) Ck45M steel (TMP-DQ). The chemical composition determined by spark optical emission spectroscopy is 0.440C-0.667Mn-0.294Si-0.009P-0.036S-0.218Cr-0.221Cu-0.0002B-0.034Mo (weight %). A $200 \times 80 \times 60$ mm block machined from the casting was soaked at $1150^\circ$C for 1 h and thermomechanically rolled in two stages, as shown in Figure 1. Stage 1 comprised hot rolling in three passes above the no-recrystallization temperature ($T_{nr}$) to a thickness of $\sim 31.5$ mm with $\sim 0.2$ strain/pass. Stage 2 comprised controlled rolling in the $T_{nr}$ regime (below $\sim 950^\circ$C) comprising three passes ($\sim 0.2$ strain/pass) to a thickness of $15.5$ mm with the finish rolling temperature at $\sim 830^\circ$C (above the ferrite start temperature, $A_r3$). Immediately after hot rolling, the sample was DQ in water to room temperature. Tensile tests were performed according to ASTM E8M standard.

Specimen shapes used for $S-N$ and near-threshold FCGR tests are shown in Figure 2A,B, respectively. All specimens were machined with their longitudinal axis in rolling direction. The surface in the gauge section of machined specimens was ground with emery paper up to
2000 and electropolished to remove any residual stresses.

Small drilled holes were introduced into the surface of some specimens. Hole diameters, \( d \), were 50, 100, 200, and 300 μm, and the depths of holes, \( h \), were the same as the diameter; that is, \( h = d \). In addition, 2- and 3-hole defects were introduced with hole diameters of 50 μm oriented perpendicular to the loading direction; see Figure 2C.

Fully reversed tension-compression fatigue tests were performed using ultrasonic fatigue testing equipment with a cycling frequency of about 19 kHz. The vibration of one specimen’s end is measured with an induction coil and used to control cyclic loading in a closed-loop circuit; that is, ultrasonic fatigue experiments are displacement controlled. The specimen is cycled in resonance, and the resonance frequency is controlled in a second closed-loop circuit. To prevent self-heating of specimens during testing, pulsed loading was applied in addition to compressed-air cooling. A detailed description of the ultrasonic testing technique can be found in Mayer23. Tests were performed in ambient air at approximately 23°C and 50% relative humidity (laboratory environment with air conditioning and controlled humidification).

In the FCGR specimens, a single-edge notch with a length of 1 mm was introduced by electrical discharge machining, acting as a well-defined crack initiation site. Measurements were started at crack growth rates of about \( 10^{-8} \) m/cycle, and the stress intensity factor, \( \Delta K \), was decreased in steps of 3–7% (with the smaller steps in the near-threshold regime) after a crack extension of at least 50 μm was detected. This was repeated until the threshold, \( \Delta K_{th,lc} \), was reached—which was determined as the stress intensity factor range where no crack extension could be observed within \( 10^{10} \) cycles. Fatigue crack growth was observed with a digital camera and a zoom lens allowing magnifications up to 680-fold (magnifications between 420- and 680-fold were used, with higher magnification at lower FCGRs) and a resolution of approximately 1 μm. Extremely slow crack growth with rates as low as \( 2 \times 10^{-14} \) m/cycles was observed, which could be only measured due to the high testing frequency (in the near-threshold regime, continuous loading at approximately 19 kHz was applied). It is noted that
accumulating $10^{10}$ cycles nevertheless takes more than 6 days, resulting in a measurement time of several weeks for the determined FCGR curve.

3 | RESULTS

3.1 | Microstructural and static mechanical property evaluation

Figure 3A presents a typical field-emission scanning electron micrograph (FE-SEM) of the investigated TMP-DQ processed steel after etching. The microstructure consists mainly of fine packets and blocks of martensite. Straining of austenite prior to quenching resulted in refinement and randomization of packets and blocks of martensite. Untransformed, so-called retained austenite (RA) present as films between martensitic laths is difficult to discern from FE-SEM micrographs. Thin film examinations in transmission electron microscope (TEM) including bright and dark field imaging clearly revealed that the material contains highly dislocated lath martensite structure separated by fine RA films (Figure 3B). The dark field image (Figure 3C) confirms the presence of film-like RA. The volume fractions of martensite and RA are estimated 96% and 4%, respectively, by X-ray diffraction analysis corroborated well by TEM results. Furthermore, energy-dispersive X-ray spectroscopy (EDX) of the micrograph shown in Figure 3A revealed presence of aluminum/calcium oxide (Al$_2$O$_3$–CaO) and manganese sulfide (MnS) inclusions with circular and elongated morphology, respectively.

The experimentally determined high yield strength (0.2% proof stress of $\sim$1613 MPa) in combination with ultrahigh tensile strength ($\sim$2439 MPa) and reasonable total elongation ($\sim$3%) corroborate the observed microstructure as shown in Figure 3A–C. The nanolath martensitic matrix imparts the high strength, whereas the finely divided, film-like interlath RA provides the work hardening capacity. Furthermore, the high hardness (717 HV10) correlates well with the determined tensile strength.

3.2 | S–N tests with smooth specimens

Fatigue test results with smooth S–N specimens are plotted in Figure 4. Tests were performed at stress amplitudes between $\sigma_a = 570$ and 700 MPa. There is a huge scatter...
of fatigue lifetimes at each stress amplitude, up to five orders of magnitude. No fatigue limit could be determined, and failure even occurred after $2.56 \times 10^{10}$ cycles. Two specimens were stopped after more than $1 \times 10^{10}$ cycles without failure (marked with arrows in Figure 4).

FE-SEM images of typical fracture surfaces are shown in Figure 5. Failure mainly occurred from nonmetallic inclusions located at the surface or in the interior (Figure 5A,B) that were identified as Al$_2$O$_3$–CaO with EDX. For some specimens, the cause of failure could not be identified (Figure 5C,D, marked as “unknown” in Figure 4). EDX analyses revealed no differences in chemical composition at the crack initiation location and the matrix. The high-magnification image in Figure 5C shows both the specimen surface (top) and the fracture surface (bottom), which was achieved by tilting the specimen in the SEM. A structure similar to bainite is visible at the surface which might be the source of failure (see, e.g., the investigations on bearing steels by Murakami$^1$). Further, some specimens failed from surface pits—as will be shown in Section 4.3—that probably originated from inclusions during electropolishing of specimens. From a fracture mechanics point of view, small pits can be treated similarly to surface inclusions,$^1$ and no further differentiation between pits and surface inclusions will be made in the following.

One specimen was polished with 1-μm diamond slurry before testing at $\sigma_a = 625$ MPa in order to achieve a mirror-like surface. After $N = 1.20 \times 10^{10}$ cycles, the surface in the gauge section was examined with a light microscope. Some locations with crack-like features were found around nonmetallic inclusions. The specimen was afterward tested at $\sigma_a = 650$ MPa for another $N = 1.00 \times 10^{10}$ cycles, and the same locations were reinvestigated. A slight increase in crack length could be observed which proofs the existence of nonpropagating cracks. Examples of nonpropagating cracks at an inclusion are shown in Figure 6A.

### 3.3 S–N tests with specimens containing drilled holes

Fatigue tests with specimens containing drilled holes were performed up to $10^9$ cycles, and the results are plotted in Figure 7. The $S$–$N$ curves exhibit clear knee-points between $10^5$ and $10^6$ cycles. Only one specimen failed due to crack initiation at a drilled hole in the VHCF regime ($d = 100 \mu$m, $\sigma_a = 575$ MPa, $N_f = 7.47 \times 10^8$ cycles). Another specimen with a 1-hole defect ($d = 50 \mu$m) failed after $N_f = 6.06 \times 10^8$ cycles at $\sigma_a = 625$ MPa, but fracture originated from another

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**FIGURE 5** Origins of fatigue fracture: (A) surface inclusion, $\sigma_a = 625$ MPa, $N_f = 2.34 \times 10^5$ cycles; (B) interior inclusion, $\sigma_a = 625$ MPa, $N_f = 2.59 \times 10^{10}$ cycles; (C) surface without inclusion, $\sigma_a = 620$ MPa, $N_f = 3.27 \times 10^5$ cycles; and (D) interior without inclusion, $\sigma_a = 675$ MPa, $N_f = 5.67 \times 10^4$ cycles
location at the surface rather than from the drilled hole. Therefore, this specimen is marked as runout in Figure 7.

The runout specimens were observed for non-propagating cracks after fatigue testing. However, due to a slight etching of the specimens’ surface that occurred during electropolishing, it was in some cases difficult to decide whether a crack was present or not. Therefore, some runout specimens were polished with 1-μm diamond slurry and fatigue loaded for another \(10^7\) cycles. Clear evidence of a nonpropagating crack could be only provided at the edge of a 1-hole defect with \(d = 100\) μm; see Figure 6B.

3.4 | Near-threshold fatigue crack growth tests

The FCGR test results are shown in Figure 8. The stress intensity factor ranges were calculated according to following equation\(^{24}:\)
\[ \Delta K = \Delta \sigma \cdot \sqrt{\pi \cdot a \cdot Y_v \left( \frac{a}{W} \right)} \]  \tag{1}

with \( Y_v = 1.12 - 0.04 \cdot (a/W) + 0.53 \cdot (a/W)^2 + 7.8 \cdot (a/W)^3 - 24.0 \cdot (a/W)^4 \) that was deduced from finite element calculations for single-edge notched ultrasonic-fatigue specimens as shown in Figure 2B.\(^{25}\) In Equation 1, \( \Delta \sigma \) is the remote stress range (calculated from the measured strain range multiplied with the Young's modulus), \( a \) is the crack length (including notch depth), and \( W \) is the specimen width (\( W = 15 \text{ mm} \)).

Crack growth rates between \( 3 \times 10^{-8} \) and \( 2 \times 10^{-14} \) m/cycles were observed, which means that propagation rates significantly below one Burgers vector per load cycle could be measured. Data points marked with arrows indicate measurements where no crack propagation could be observed within \( 10^{10} \) cycles. The mean value of the determined long-crack threshold-stress intensity factor range is \( \Delta K_{th,lc} = 2.4 \text{ MPa} \sqrt{\text{m}} \). Note that the whole stress range (tension and compression part) was used to calculate the stress intensity factor range at \( R = -1 \); that is, the threshold-stress intensity factor amplitude as well as the maximum stress intensity factor at the threshold is \( 1.2 \text{ MPa} \sqrt{\text{m}} \). In order to verify crack growth at extremely low propagation rates and the low threshold stress intensity factor range, three specimens were tested (marked with different symbols in Figure 8). The curves measured with different specimens are in good accordance with each other; hence, it can be assumed that the determined results are valid.

### DISCUSSION

#### 4.1 Observation of nonpropagating cracks

Nonpropagating cracks with lengths of approximately 10 \( \mu \text{m} \) could be observed at both nonmetallic surface inclusions and small drilled holes (as shown in Figure 6). This is a surprising observation since self-arrested cracks are rarely observed in ultrahigh-strength steels.\(^{26,27}\) The presence of nonpropagating cracks after cyclic loading at subcritical loads (i.e., at or below the fatigue limit stress amplitude) is a clear indication that fracture mechanics approaches might be used to predict the fatigue strength in the presence of small defects. This is due to the fact that the fatigue limit is determined by the critical condition for crack propagation rather than for crack initiation, which means that the fatigue limit can be correlated with the threshold-stress intensity.

As demonstrated by Nisitani\(^{28}\), the root radius of a notch or a defect, \( \rho \), influences the fatigue limit if the value of \( \rho \) is above a critical value. This critical notch root radius, \( \rho_0 \), depends on the material. If \( \rho \leq \rho_0 \), the fatigue limit is determined by the threshold condition for crack propagation—which means that nonpropagating cracks are observable (see Figure 9). A notch with a small crack at its root may be regarded equivalent to a crack\(^1\) (i.e., at the crack tip, \( \rho \) is virtually zero—indeed of the root radius of the initial notch). Thus, fracture mechanics principles are applicable. If \( \rho > \rho_0 \), the threshold condition for crack initiation determines the fatigue limit. In this case, notch-fatigue concepts may serve to predict the fatigue strength rather than a fracture mechanics approach.
approach. For example, the relative stress gradient approach by Siebel and Stieler\textsuperscript{29} has been successfully applied for martensitic stainless steels.\textsuperscript{4} However, the applicability of this approach to small defects needs to be further investigated.

Values of $\rho_0$ typically vary between 100 and 500 $\mu$m— with a tendency of decreasing values of $\rho_0$ with increasing tensile strength.\textsuperscript{28} However, a systematic investigation of martensitic stainless steels with tensile strengths up to 1 GPa\textsuperscript{4,30} has shown that $\rho_0$ can be as low as 25–100 $\mu$m. Therefore, it might be assumed that the critical defect size for steels with tensile strengths above 2 GPa is even smaller and that small drilled holes (and maybe even nonmetallic inclusions) rather behave like blunt notches.\textsuperscript{26} The observation of a nonpropagating crack at a 100-$\mu$m hole as shown in Figure 6B, however, confutes this assumption (although it is conceivable that local microstructural inhomogeneities may act similarly to small, sharp notches) and suggests that fracture mechanics principles can be applied—as will be discussed in the following sections.

### 4.2 Fracture mechanics evaluation

The stress intensity factor ranges, $\Delta K$, for nonmetallic inclusions and drilled holes were calculated by the widely used equation proposed by Murakami\textsuperscript{1}:

$$\Delta K = f \cdot \Delta \sigma \cdot \sqrt{\pi \cdot \sqrt{\text{area}}}$$ \hspace{1cm} (2)

with $f = 0.65$ for surface defects and $f = 0.5$ for interior defects. Equation 2 gives the maximum value of $\Delta K$ along the front of the defect in MPa$\sqrt{\text{m}}$ if the cyclic stress range, $\Delta \sigma$, is in MPa and $\sqrt{\text{area}}$ is in m. The size parameter $\sqrt{\text{area}}$ introduced by Murakami and Endo\textsuperscript{31} is the square root of the projection area of the defect perpendicular to the loading direction, and values for $\sqrt{\text{area}}$ were determined from FE-SEM fractographs as shown in Figure 5. Sizes of $\sqrt{\text{area}} = 13–116\mu$m were determined for nonmetallic inclusion and $\sqrt{\text{area}} = 10–56\mu$m for surface pits. Runout specimens were retested at a higher stress amplitude to enable determination of the crack-initiating defect’s size.

In Figure 10, the calculated values of $\Delta K$ are plotted versus the number of cycles to failure. Failure from surface inclusions (or pits) and interior inclusions is represented by open and solid symbols, respectively (Figure 10A). Further, interior inclusions with sizes of $\sqrt{\text{area}} < 60\mu$m (gray symbols) and $\sqrt{\text{area}} > 60\mu$m (black symbols) are separately plotted. With this classification—and in contrast to the $S-N$ curve shown in Figure 4—clear fatigue-lifetime curves could be plotted. The separation of small and large interior inclusions indicates that (for a given stress intensity factor range) fatigue lifetimes increase when the inclusion sizes become larger. This might be a hint that the threshold-stress intensity factor range, $\Delta K_{\text{th}}$, is size dependent (in the nonmetallic inclusions size range).

In contrast, the fatigue limits determined with specimens containing drilled holes can be related to a constant stress intensity factor range of around $\Delta K_{\text{th}} \approx 12\text{ MPa}\sqrt{\text{m}}$.
as shown in Figure 10B; 1-hole defects with hole diameters of \( d = 200 \) and 300 \( \mu \text{m} \) exhibit almost the same \( \Delta K-N \) curves, but fatigue lifetimes are shorter for 2- and 3-hole defects when \( \Delta K \) is above the threshold. This can be explained by the significantly smaller notch root radius of \( \rho = d/2 = 25 \mu \text{m} \) for the latter, which results in higher stress concentration factors. However, it is important to notice that although the stress concentration factor may influence the fatigue lifetimes, the fatigue limit is rather determined by the stress intensity factor.

Murakami and Endo\textsuperscript{32} proposed a simple model to predict the size-dependent threshold-stress intensity factor range for small cracks and defects, which has been successfully applied to several materials.\textsuperscript{1} Beside the size parameter \( \sqrt{\text{area}} \)—which is eponymous for the model—only the Vickers hardness, \( HV \), must be known to calculate the threshold-stress intensity factor range:\textsuperscript{3}

\[
\Delta K_{th} = g \cdot (HV + 120) \cdot (\sqrt{\text{area}})^{1/3} \tag{3}
\]

with \( g = 3.3 \times 10^{-3} \) for surface cracks/defects and \( g = 2.77 \times 10^{-3} \) for interior cracks/defects. In Equation 3, \( \Delta K_{th} \) is in MPa\( \sqrt{\text{m}} \), \( \sqrt{\text{area}} \) in \( \mu \text{m} \), and \( HV \) in kgf/mm\(^2\).

The applicability of Equation 3 is limited to small defects; that is, the threshold-stress intensity factor should be below the constant long-crack threshold, \( \Delta K_{th,lc} \). From tests with specimens containing small drilled holes (Figure 7), a constant threshold of \( \sim 12 \text{ MPa} \sqrt{\text{m}} \) can be assumed (Figure 10B). However, this value is significantly larger than the long-crack threshold of \( \Delta K_{th,lc} = 2.4 \text{ MPa} \sqrt{\text{m}} \) determined by fatigue crack growth measurements in ambient air (Figure 8). Deviations from \( \Delta K_{th,lc} \) for large defects have been reported for martensitic stainless steels\textsuperscript{4} and spheroidal cast iron,\textsuperscript{33} but they are mainly found at load ratios different from \( R = -1 \) and could be explained by the influence of mean stress on the crack tip constraints.\textsuperscript{4}

The extremely slow crack growth rates measured in the near-threshold regime as shown in Figure 8 rather suggest environmental influences that enable crack growth at extremely low stress intensity factors. Also, the value of \( 2.4 \text{ MPa} \sqrt{\text{m}} \) is in the range of the effective threshold-stress intensity factor range of steels, but crack growth at stress intensities close to \( \Delta K_{th,cr} \) should be only relevant at high load ratios—and not at fully-reversed loading. Therefore, in the following, the long-crack threshold, \( \Delta K_{th,lc} \), will be distinguished from the threshold-stress intensity factor range for large defects, \( \Delta K_{th,ld} \), which is according to Figure 10B approximately \( \Delta K_{th,ld} = 12 \text{ MPa} \sqrt{\text{m}} \). The transition size between small and large defects, \( \sqrt{\text{area}}_{\text{trans}} \), can be hence calculated by following equation\textsuperscript{30}:

\[
\sqrt{\text{area}}_{\text{trans}} = \left( \frac{\Delta K_{th,ld}}{g \cdot (HV + 120)} \right)^{3/2} \tag{4}
\]

Transition sizes of \( \sqrt{\text{area}}_{\text{trans,}s} = 82 \mu \text{m} \) for surface defects and \( \sqrt{\text{area}}_{\text{trans,i}} = 139 \mu \text{m} \) for interior defects are calculated for the investigated ultrahigh-strength steel. Similar transitions sizes were determined for martensitic stainless steels as reported in Schönbauer et al.\textsuperscript{4,30} Comparison between the transition sizes for surface and interior defects—as well as the stress intensity factor ranges according to Equation 2—reveals that the size of an interior defect (in \( \sqrt{\text{area}} \)) must be by a factor of 1.69 larger to be equally detrimental than a surface defect; see also Murakami.\textsuperscript{1}

In Figure 11, the stress intensity factor range, \( \Delta K \), is plotted versus the defect size, \( \sqrt{\text{area}} \). Solid lines mark the threshold values according to Equation 3 and \( \Delta K_{th,ld} \), respectively. Since \( \Delta K_{th,ld} \) was determined from the experimental results, all failed specimens containing drilled holes with sizes of \( \sqrt{\text{area}} > \sqrt{\text{area}}_{\text{trans,}s} \) lie per definition above the prediction line (Figure 11A). But, surface inclusions and pits were all smaller than the transition size of 82 \( \mu \text{m} \), and hence, the applicability of Equation 3 needs to be verified. For specimens that failed below one million cycles (large, solid circles), the \( \sqrt{\text{area}} \)-parameter model is accurate within \( \pm 10\% \). If failure occurred after \( 10^6 \) cycles, however, the prediction becomes nonconservative. The stress intensity factor for the specimen that failed from a 10-\( \mu \text{m} \) pit, for example, was 27\% below the predicted threshold value. For interior inclusions, the prediction according to Equation 3 also slightly overestimates the threshold values; see Figure 11B. The specimen that fractured after \( 2.59 \times 10^{10} \) cycles (Figure 5B) failed 13\% below the calculated fatigue strength.

By substituting the stress intensity factor range and the cyclic stress range in Equation 2 by its threshold, \( \Delta K = \Delta K_{th} \), and the fatigue limit, \( \Delta \sigma = 2 \cdot \sigma_w \), respectively, the fatigue limit stress amplitude, \( \sigma_w \), can be derived. For small cracks or defect, \( \Delta K_{th} \) is given by Equation 3, and it follows\textsuperscript{1}

\[
\sigma_w = \frac{h \cdot (HV + 120)}{(\sqrt{\text{area}})^{1/6}} \quad \text{for} \quad \sqrt{\text{area}} \leq \sqrt{\text{area}}_{\text{trans}} \tag{5}
\]

with \( h = 1.43 \) for surface cracks/defects and \( h = 1.56 \) for interior cracks/defects.
For large defect, $\Delta K_{th} = \Delta K_{th,ld}$, and the fatigue limit stress amplitude can be calculated with

$$\sigma_{w,ld} = \frac{i \cdot \Delta K_{th,ld}}{\sqrt{\text{area}}}$$

for $\sqrt{\text{area}} \geq \sqrt{\text{area}_{\text{trans}}}$ (6)

where $i = 434$ for surface defects and $i = 564$ for interior defects.

With this, the experimental results can be plotted in Kitagawa–Takahashi diagrams, as demonstrated in Figure 11C,D, which allows to correlate the fatigue limit with the defect size.

Note that Equations 3, 5, and 6 are only applicable for $R = -1$. In the presence of mean stresses, a term accounting for the load ratio must be included as described in Schönbauer et al.\textsuperscript{1,30} For martensitic stainless steels, the transition size (Equation 4) was found to be independent of load ratio.\textsuperscript{4} Although it is expected that the mean load sensitivity increases with tensile strength, systematic investigations with ultrahigh strength steels at different load ratios are necessary to give further insight.

### 4.3 VHCF failure from surface defects

In steels, fatigue failure from the surface typically occurs within $10^7$ cycles or earlier.\textsuperscript{1,34} This was also observed for the investigated steel in the presence of small drilled...
holes where—with one exception—all specimens failed before $10^6$ or survived $10^9$ cycles. Surface failure in the VHCF regime is typically attributed to environmental effects. In ambient air, this is mainly an issue for high-strength steels (see, e.g., Nishimura et al.\textsuperscript{35}) since corrosion resistance tendentially decreases with increasing tensile strength.

In the Kitagawa–Takahashi diagram plotted in Figure 12, a short-dashed, red line denotes the stress amplitudes above which nonpropagating cracks could be observed. Symbols with centerlines mark the two specimens with nonpropagating cracks that are shown in Figure 6. The size of the nonmetallic inclusion shown in Figure 6 was estimated as $\sqrt{\text{area}} = 10 \, \mu\text{m}$ (assuming a semi-circular shape). In addition, a dashed, red line represents the experimentally determined long-crack threshold-stress intensity factor of $\Delta K_{\text{th},lc} = 2.4 \, \text{MPa}\sqrt{\text{m}}$. Above the black, solid prediction line—representing Equations 5 and 6—fatigue failure is expected, while no failure should occur if applied stress amplitudes are below this line. The horizontal part of the prediction line—which declines at a crack length of $1.3 \, \mu\text{m}$—marks the upper bound of the fatigue strength which can be estimated (for steel) by the simple equation $\sigma_w = 1.6 \times HV$.\textsuperscript{1} A possible explanation for the observed failure in the hatched area below the predicted fatigue limit in Figure 12 (small solid circles) will be given in the following:

At small, sharp defects such as nonmetallic inclusions, cracks can initiate at stress amplitudes even significantly below the fatigue limit. These cracks would arrest in a benign environment due to the stress gradient and the built-up of crack closure mechanisms.\textsuperscript{36–38} However, in a chemically active environment—and ambient air might be assumed as a weak corrosive atmosphere for ultrahigh-strength steels—the crack growth threshold is decreased resulting in a lowering of the original fatigue limit.\textsuperscript{20} In other words, the gap between the crack initiation and the crack propagation limit, which are appreciably different in noncorrosive environments, becomes smaller. This assumption is strongly supported by the extremely low near-threshold FCGRs and threshold-stress intensity factor measured in ambient air (Figure 8). The corrosive influence of ambient air becomes especially important for surface defects with sizes below 30 $\mu\text{m}$. In the presence of comparably large drilled holes, where the threshold-stress intensity factor is size independent and the notch root is relatively large, the crack-initiation limit is only slightly below the propagation limit. The corrosive atmosphere—which is too weak to decrease the crack-initiation limit, for example, by pitting—therefore, may not cause a significant reduction in fatigue strength.

Furthermore, DQ ultrahigh-strength steels are highly susceptible to hydrogen embrittlement\textsuperscript{39} due to high residual stresses exerted by the large fraction of martensite in the final microstructure.\textsuperscript{40} Hydrogen can be trapped at different sites such as dislocations, grain boundaries, precipitates, or inclusions. Depending on the activation energy, weak or reversible hydrogen traps must be distinguished from irreversible traps where a significant amount of hydrogen can be captured without considerable impact on cracking.\textsuperscript{41} Reversible traps, in contrast, may act as a supplier of diffusible hydrogen to the crack tip.\textsuperscript{42} The observation of intergranular crack paths supports the assumption that hydrogen embrittlement affects fatigue crack growth. However, further investigations on the effects of environment and hydrogen on the near-threshold fatigue crack growth are necessary to give more insight.

The fracture surface of the specimen that failed from a small pit in the VHCF regime, at a stress amplitude 26% below the predicted fatigue limit according to Equation 5, is shown in Figure 13A. Transgranular, cleavage-like facets in the near vicinity of the pit with a nonplanar crack path are visible. Similar—but intergranular—facets are observable adjacent to the crack-initiating inclusion shown in Figure 13B. Such near-threshold facets are expected to be preliminarily environmentally induced\textsuperscript{43,44} and can be, for example, observed when fatigue cracks initiate at corrosion pits.\textsuperscript{45,46} It is further well-known that dissolved hydrogen benefits intergranular cracking.\textsuperscript{47,48} It might be considered that these inter- and transgranular facets can be treated as effective flaws.

**FIGURE 12** Kitagawa–Takahashi diagram for surface defects. Specimens with nonpropagating cracks (shown in Figure 6) are highlighted by red symbols with centerlines [Colour figure can be viewed at wileyonlinelibrary.com]
The sizes of the resulting effective defects are marked in Figure 13, and these effective defect sizes were used to calculate the fatigue limits according to Equation 5. Plotting the applied stress amplitudes normalized by the predicted fatigue limits (Equations 5 and 6) versus the number of cycles to failure, as shown in Figure 14, reveals that an increase in the effective defect size serves well to explain fatigue failure in the VHCF regime from inclusions and pits that would be nondetrimental under benign environmental conditions.

4.4 Fatigue limit in the presence of interior defects

The normalized stress amplitudes versus number of cycles to failure for interior nonmetallic inclusions are plotted in Figure 15. Compared to Figure 10, where ΔK was plotted versus the number of cycles, scatter in fatigue data could be significantly reduced. This is due to the fact that not only the size of inclusions but also the size-dependent threshold-stress intensity factor range (Equation 3) is considered. Several examples of normalized S–N curves have been published showing a similar reduction in scatter.\textsuperscript{9,49–51} However, there is still a tendency of longer fatigue lifetimes for larger inclusions noticeable in Figure 15. This is similar to specimens containing drilled holes where fatigue lifetimes are shorter when the holes are smaller (and/or the notch root radius is reduced), see Figures 10B and 14, and might be explained by differences in stress distribution at small and large inclusions.

In the case of an accurate prediction of σ\textsubscript{w}, no failure should occur below σ\textsubscript{a}/σ\textsubscript{w} = 1. In the VHCF regime, however, the \sqrt{area}-parameter model delivers slightly nonconservative predictions, as shown in Figure 15. For example, fracture occurred at a stress amplitude 13% below the fatigue limit calculated with Equation 5 (i.e., at σ\textsubscript{a}/σ\textsubscript{w} = 0.87), however, after an extremely high number of cycles of more than 2 \times 10\textsuperscript{10}.
Therefore, the prediction might be considered as acceptably accurate. Nevertheless, assuming that the explanation given in the last section (i.e., that the fatigue limit is reduced by an environmental interaction and dissolved hydrogen, and Equation 5 generally serves well to predict the fatigue limit of the investigated steel in the presence of surface defects) is correct, there must be a reason for the observed reduction in fatigue strength in the case of internal fracture in the VHCF regime. Also, similar (nonconservative) results are reported for other ultrahigh-strength steels, although Equation 5 is in good agreement with experimentally determined fatigue strengths in the VHCF regime for high-strength steels (with tensile strength below 2 GPa), even in the case of failure from internal inclusions.

Fatigue fracture from the interior forms a so-called fish eye morphology, and a fine-grained area is often observed around the crack-initiation location if failure occurs in the VHCF regime. This area is mostly called “optically dark area” (ODA) since it appears dark when observed with an optical microscope—but bright with an SEM—or “fine granular area” due to its morphology and can be seen in Figure 5B,C. Hydrogen trapped by nonmetallic inclusions can significantly influence the formation of ODAs as reported by Murakami et al. and the concomitant, localized embrittlement may explain the reduction of the fatigue strength as displayed in Figure 15 (i.e., similar effects as expected for surface inclusions might be expected). But, ODAs can also be observed in the absence of inclusions, as shown in Figure 5C, as long as crack initiation is from the interior. In contrast, ODAs have never been observed at surface inclusions—as long as tests were performed in ambient environment. In high vacuum, however, fine granular appearing fracture surfaces can be produced around small surface defects and even during long-crack propagation in the near-threshold regime. Tests with specimens containing small surface defects conducted in high-vacuum and air by Ooka et al. showed that fatigue lifetimes are significantly increased in vacuum. Failure occurred even in the VHCF regime, while the S–N curves in air exhibited knee-points at around $10^7$ cycles. Nevertheless, fatigue failure in vacuum occurred at stress amplitudes below the fatigue limit determined in ambient air.

Figure 16A shows the fracture surface of the specimen that initially survived $1.10 \times 10^{10}$ cycles at 625 MPa (marked as runout specimen in Figure 15). Afterward, the specimen was retested at 700 MPa, and failure occurred after $1.23 \times 10^7$ cycles. In the vicinity of the crack-initiating inclusion, a pronounced ODA is visible (the border of the ODA is marked with arrows in Figure 16A). Such an area is not identifiable at the fracture origin of another specimen that was exclusively tested at a stress amplitude of 700 MPa, although the...
inclusion size and the number of cycles to failure were comparable; see Figure 16B. It can be therefore assumed that formation of the ODA visible in Figure 16A took place during cycling at 625 MPa. Since ODA formation consumes the vast majority of the fatigue lifetime, the mean crack growth rate inside the fine-grained area was approximately $10^{-15}$ m/cycle. This is extremely slow and 1–3 orders of magnitude below FCGRs measured indirectly by two-step tests and directly by long-crack measurements in high vacuum. Hong et al. report mean crack growth rates in ODAs between $10^{-11}$ and $10^{-12}$ m/cycle for lifetimes of $10^6–10^7$ cycles and between $10^{-12}$ and $10^{-13}$ m/cycle for lifetimes of $10^7–4\times10^8$ cycles. At fatigue lifetimes exceeding $10^9$ cycles, hence, mean growth rates of $10^{-15}$ m/cycle are conceivable. Although the border of the ODA is more difficult to distinguish, similar crack growth rates are estimable for the specimen that failed after $2.59\times10^{10}$ cycles (fracture surface shown in Figure 5B).

It is often concluded that ODA formation ends when the stress intensity factor of the crack exceeds a constant value (e.g., the long-crack threshold). As shown in Figure 11—assuming the same threshold-stress intensity factor for long cracks in vacuum as determined for large surface defects, that is, $\Delta K_{\text{th,ld}} \approx 12 \text{ MPa} \sqrt{\text{m}}$—it might be however expected that the threshold-stress intensity factor is rather size dependent (as assumed by Murakami et al.) and that the transition size according to Equation 4 is $139 \mu \text{m}$. Since the border of ODAs is difficult to distinguish in the investigated material (similarly to other materials such as 17-4PH stainless steel), it is futile to evaluate the stress intensity factors of ODAs (which might rather deliver expected outcomes than objective results). Assuming that environmental conditions of interior cracks can be accurately simulated, it seems to be more appropriate to perform FCGR measurements in high vacuum and to correlate the stress intensities of small defects and inclusions with the determined thresholds.

5 | CONCLUSIONS

Ultrasonic fatigue tests up to more than $10^{10}$ cycles were performed with ultrahigh-strength CK45M steel processed by thermomechanical rolling and subsequent direct quenching. The fracture origins of smooth specimens were mainly nonmetallic inclusions located at the surface or in the interior of specimens. Systematic investigation on the defect tolerance was conducted with specimens containing small drilled holes. Following main results were obtained:
1. The fatigue limit in the presence of surface defects larger than approximately 80 μm can be correlated with a constant threshold-stress intensity factor range of $\Delta K_{th,ld} \approx 12$ MPa√m. For smaller defects, the threshold-stress intensity factor becomes size dependent, and the $\sqrt{\text{area}}$-parameter model serves well to predict the fatigue strength in the high cycle fatigue regime. Applicability of fracture mechanics for fatigue limit prediction is supported by the observation of nonpropagating cracks in specimens cycled below the fatigue limit.

2. A long-crack threshold-stress intensity factor range of $\Delta K_{th,lc} = 2.4$ MPa√m was determined in ambient air by near-threshold FCGR measurements at $R = -1$. Propagation rates as low as $2 \times 10^{-14}$ m/cycles were measured. Crack growth at extremely low propagation rates and the low threshold value were associated with environmental effects and dissolved hydrogen.

3. VHCF failure from small surface defects (<30 μm) occasionally occurred at stress amplitudes below the fatigue limit predicted by the $\sqrt{\text{area}}$-parameter model. Crack growth at stress intensities below the crack length-dependent threshold-stress intensity factor is fostered by the corrosive influence of ambient air and hydrogen embrittlement.

4. Failure from the interior occurred even beyond $2 \times 10^{10}$ cycles, indicating mean crack growth rates of approximately $10^{-15}$ m/cycles within the ODA that is formed around interior inclusions. It is assumed that the extremely low growth rates are associated with crack propagation under high-vacuum condition.

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AUTHOR CONTRIBUTIONS
Bernd M. Schönbaumer performed the conceptualization (lead), funding acquisition (equal), investigation (lead), validation (equal), methodology (lead), formal analysis (lead), supervision (lead), project administration (equal), writing—original draft preparation (lead), visualization (lead), and review and editing (equal). Sumit Ghosh performed the investigation (supporting), writing—original draft preparation (supporting), visualization (supporting), and review and editing (equal). Jukka Kömi performed the resources (equal), funding acquisition (equal), and review and editing (equal). Tero Frondelius performed the project administration (equal) and review and editing (equal). Herwig Mayer performed the validation (equal), resources (equal), and review and editing (equal).

DATA AVAILABILITY STATEMENT
The data that support the finding of this study are included within the paper.

NOMENCLATURE

| Symbol | Description |
|--------|-------------|
| $a$ | half-width of defect or crack length at surface |
| $\sqrt{\text{area}}$ | square root of the projection area of a defect perpendicular to the loading direction |
| $\sqrt{\text{area}}_{\text{trans}}$ | transition size between small and large surface defects |
| $d$ | hole diameter |
| $h$ | depth of defect |
| $HV$ | Vickers hardness |
| $N$ | number of cycles |
| $N_f$ | number of cycles to failure |
| $R$ | load ratio |
| $W$ | specimen width |
| $\Delta K$ | stress intensity factor range |
| $\Delta K_{th}$ | threshold stress intensity factor range for small cracks or defects |
| $\Delta K_{th,eff}$ | effective threshold stress intensity factor range |
| $\Delta K_{th,lc}$ | threshold stress intensity factor range for long cracks |
| $\Delta K_{th,ld}$ | threshold stress intensity factor range for large defects |
| $\Delta \sigma$ | stress range |
| $\rho$ | notch root radius |
| $\rho_0$ | critical notch root radius |
| $\sigma_a$ | stress amplitude ($\sigma_a = \Delta \sigma / 2$) |
| $\sigma_w$ | fatigue limit stress amplitude at fully-reversed loading ($R = -1$) |
| $\sigma_{w,ld}$ | fatigue limit stress amplitude in the presence of large defects |

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REFERENCES
1. Murakami Y. Metal fatigue: Effects of small defects and nonmetallic inclusions. 2nd ed. Cambridge, MA: Academic Press; 2019.
2. Nishimura Y, Endo M, Yanase K, Ikeda Y, Miyakawa S, Miyamoto N. High cycle fatigue strength of spring steel with small scratches. Proceedings of Advances in Structural Integrity, Springer, Singapore. 2018:541-548.
3. Yang ZG, Zhang JM, Li SX, et al. On the critical inclusion size of high strength steels under ultra-high cycle fatigue. Mater Sci Eng A. 2006;427(1-2):167-174. https://doi.org/10.1016/j.msea.2006.04.068

4. Schönbauer BM, Mayer H. Effect of small defects on the fatigue strength of martensitic stainless steels. Int J Fatigue. 2019;127:362-375. https://doi.org/10.1016/j.ijfatigue.2019.06.021

5. Gumbel EJ. Statistics of Extremes. Columbia University Press; 1958.

6. Karr U, Schönbauer B, Fitzka M, et al. Inclusion initiated fracture under cyclic torsion very high cycle fatigue at different load ratios. Int J Fatigue. 2019;122:199-207. https://doi.org/10.1016/j.ijfatigue.2019.01.015

7. Karr U, Sandaïji Y, Tanegashima R, et al. Inclusion initiated fracture in spring steel under axial and torsion very high cycle fatigue loading at different load ratios. Int J Fatigue. 2020;143:105525. https://doi.org/10.1016/j.ijfatigue.2020.105525

8. Chapetti MD. A simple model to predict the very high cycle fatigue resistance of steels. Int J Fatigue. 2011;33(7):833-841. https://doi.org/10.1016/j.ijfatigue.2010.12.010

9. Mayer H, Haydn W, Schuller R, Issler S, Furtner B, Bacher-Melde H. Very high cycle fatigue properties of baunitic high carbon–chromium steel. Int J Fatigue. 2009;31(2):242-249. https://doi.org/10.1016/j.ijfatigue.2008.09.001

10. Furuya Y, Hirukawa H, Kimura T, Hayashi M. Gigacycle fatigue properties of high-strength steels according to inclusion and ODA sizes. Metall Mater Trans A. 2007;38(8):1722-1730. https://doi.org/10.1007/s11661-007-9225-3

11. Matsunaga H, Sun C, Hong Y, Murakami Y. Dominant factors for very-high-cycle fatigue of high-strength steels and a new design method for components. Fatigue Fract Eng M. 2015;38(11):1274-1284. https://doi.org/10.1111/ffe.12331

12. Mayer H, Schuller R, Fitzka M, Tran D, Pennings B. Very high cycle fatigue of nitrided 18Ni maraging steel sheet. Int J Fatigue. 2014;64:140-146. https://doi.org/10.1016/j.ijfatigue.2014.02.003

13. Akinbiyi J, Miyamoto N, Tsuru H, Tanaka K. Notch effect on fatigue strength reduction of bearing steel in the very high cycle regime. Int J Fatigue. 2006;28(11):1555-1565. https://doi.org/10.1016/j.ijfatigue.2005.04.017

14. Murakami Y, Nomoto T, Ueda T, Murakami Y. On the mechanism of fatigue failure in the superlong life regime (N>107 cycles). Part 1: influence of hydrogen trapped by inclusions. Fatigue Fract Eng Mater Struct. 2000;23(11):893-902. https://doi.org/10.1046/j.1460-2695.2000.00328.x

15. Sakai T, Sato Y, Oguma N. Characteristic S–N properties of high-carbon–chromium-bearing steel under axial loading in long-life fatigue. Fatigue Fract Eng Mater Struct. 2002;25(8–9):765-773. https://doi.org/10.1046/j.1460-2695.2002.00574.x

16. Shiozawa K, Morii Y, Nishino S, Lu L. Subsurface crack initiation and propagation mechanism in high-strength steel in a very high cycle fatigue regime. Int J Fatigue. 2006;28(11):1521-1532. https://doi.org/10.1016/j.ijfatigue.2005.08.015

17. Nakamura T, Oguma H, Shinohara Y. The effect of vacuum-like environment inside sub-surface fatigue crack on the formation of ODA fracture surface in high strength steel. Proc Eng. 2010;2(1):2121-2129. https://doi.org/10.1016/j.proeng.2010.03.228

18. Hong Y, Liu X, Lei Z, Sun C. The formation mechanism of characteristic region at crack initiation for very-high-cycle fatigue of high-strength steels. Int J Fatigue. 2016;89:108-118. https://doi.org/10.1016/j.ijfatigue.2015.11.029

19. Grad P, Reuscher B, Brodiansky A, Kopnarski M, Kerscher E. Mechanism of fatigue crack initiation and propagation in the very high cycle fatigue regime of high-strength steels. Scr Mater. 2012;67(10):838-841. https://doi.org/10.1016/j.scriptamat.2012.07.049

20. Miller KJ, O’Donnell WJ. The fatigue limit and its elimination. Fatigue Fract Eng Mater Struct. 1999;22(7):545-557. https://doi.org/10.1046/j.1460-2695.1999.00204.x

21. Furuya Y, Hirukawa H, Takeuchi E. Gigacycle fatigue in high strength steels. Sci Technol Adv Mater. 2019;20(1):643-656. https://doi.org/10.1080/14686996.2019.1610904

22. Hong Y, Zhao A, Qian G, Zhou C. Fatigue strength and crack initiation mechanism of very-high-cycle fatigue for low alloy steels. Metall Mater Trans A. 2012;43(8):2753-2762. https://doi.org/10.11161-011-0816-7

23. Mayer H. Recent developments in ultrasonic fatigue. Fatigue Fract Eng Mater Struct. 2016;39(1):3-29. https://doi.org/10.1016/j.ffe.2015.11.011

24. Mayer H. Fatigue crack growth and threshold measurements at very high frequencies. Int Mater Rev. 1999;44(1):1-34. https://doi.org/10.1179/imm.1999.44.1.1

25. Mayer HR, Stanzl-Tschegg SE, Tan DM. FEM modelling of stress intensity factors for fatigue crack growth at ultrasonic frequencies. Eng Fract Mech. 1993;45(4):487-495. https://doi.org/10.1016/0013-7944(93)00255-q

26. Áman M, Wada K, Matsunaga H, Remes H, Marquis G. The influence of interacting small defects on the fatigue limits of a pure iron and a bearing steel. Int J Fatigue. 2020;135:105560. https://doi.org/10.1016/j.ijfatigue.2020.105560

27. Wada K, Abass A, Okazaki S, Fukushima Y, Matsunaga H, Tsuzuki K. Fatigue crack threshold of bearing steel at a very low stress ratio. Proc Struct Integr. 2017;7:391-398. https://doi.org/10.1016/j.prostr.2017.11.104

28. Nisitani H. Effects of size on the fatigue limit and the branch point in rotary bending tests of carbon steel specimens. Bull JSME. 1968;11(48):947-957. https://doi.org/10.1299/jsme1958.11.947

29. Siebel E, Stieler M. Ungleichförmige Spannungsverteilung bei schwingender Beanspruchung. VDI-Z. 1955;97:121-126.

30. Schönbauer BM, Yanase K, Endo M. The influence of various types of small defects on the fatigue limit of precipitation-hardened 17-4PH stainless steel. Theor Appl Fract Mech. 2017;87:35-49. https://doi.org/10.1016/j.tafmec.2016.10.003

31. Murakami Y, Endo M. Quantitative evaluation of fatigue strength of metals containing various small defects or cracks. Eng Fract Mech. 1983;17:1-15. https://doi.org/10.1016/0013-7944(83)90018-8

32. Murakami Y, Endo M. Effect of hardness and crack geometries on ΔKth of small cracks emanating from small defect. In: Miller KJ, de Los Rios ER, eds. The Behavior of Short Fatigue Crack. Vol EGF Pub. 1. London: Mechanical Engineering Publications; 1986:275-293.

33. Yamabe J, Kobayashi M. Effect of hardness and stress ratio on threshold stress intensity factor ranges for small cracks and...
long cracks in spheroidal cast irons. J Solid Mech Mater Eng. 2007;1(5):667-678. https://doi.org/10.1299/jmmpe.1667
34. Jiang Q, Sun C, Liu X, Hong Y. Very-high-cycle fatigue behav-
or of a structural steel with and without induced surface defects. Int J Fatigue. 2016;93:352-362. https://doi.org/10.1016/j.
jifatigue.2016.05.032
35. Nishimura Y, Yanase K, Ikeda Y, et al. Fatigue strength of spring steel with small scratches. Fatigue Fract Eng Mater Struct. 2018;41(7):1514-1528. https://doi.org/10.1111/fle.12793
36. Wolf E. Fatigue-crack closure under cyclic tension. Eng Fract Mech. 1970;2(1):37-45. https://doi.org/10.1016/0013-7944(70)90028-7
37. Suresh S, Ritchie RO. Near-threshold fatigue crack propagation: a perspective on the role of crack closure. In: Fatigue Crack Growth Threshold Concepts; 3–5 Oct. 1983. Philadelphia, PA; 1983.
38. Pippin R, Hohenwarter A. Fatigue crack closure: a review of the physical phenomena. Fatigue Fract Eng Mater Struct. 2017;40(4):471-495. https://doi.org/10.1111/fle.12578
39. Garrison WM, Moody NR. 12 - Hydrogen embrittlement of high strength steels. In: Gangloff RP, Somerday BP, eds. Gaseous Hydrogen Embrittlement of Materials in Energy Technologies. Vol. 2. Woodhead Publishing; 2012:421-492.
40. Venezuela J, Liu Q, Zhang M, Zhou Q, Atrens A. A review of hydrogen embrittlement of martensitic advanced high-strength steels. Corrosion Rev. 2016;34(3):153-186. https://doi.org/10.1515/correv-2016-0006
41. Turnbull A. 4 - Hydrogen diffusion and trapping in metals. In: Gangloff RP, Somerday BP, eds. Gaseous Hydrogen Embrittlement of Materials in Energy Technologies. Vol.1. Woodhead Publishing; 2012:89-128.
42. Thomas RLS, Scully JR, Gangloff RP. Internal hydrogen embrittlement of ultrahigh-strength AERMET 100 steel. Metall Mater Trans A. 2003;34(2):327-344. https://doi.org/10.1007/s11661-003-0334-3
43. Ritchie RO. Near-threshold fatigue-crack propagation in steels. Int Mater Rev. 1979;24(1):205-230. https://doi.org/10.1179/imtr.1979.24.1.205
44. Schönauer BM, Stanzl-Tschegg SE. Influence of environment on the fatigue crack growth behaviour of 12% Cr steel. Ultrasonics. 2013;53(8):1399-1405. https://doi.org/10.1016/j.ultras.2013.02.007
45. Schönauer BM, Stanzl-Tschegg SE, Perlega A, et al. The influence of corrosion pits on the fatigue life of 17-4PH steel turbine blade steel. Eng Fract Mech. 2015;147:158-175. https://doi.org/10.1016/j.engfracmech.2015.08.011
46. Schönauer B, Perlega A, Karr UP, Gandy D, Stanzl-Tschegg S. Pit-to-crack transition under cyclic loading in 12% Cr steel turbine blade steel. Int J Fatigue. 2015;76:19-32. https://doi.org/10.1016/j.jfatigue.2014.10.010
47. Novak P, Yuan R, Somerday BP, Sofronis P, Ritchie RO. A statistical, physical-based, micro-mechanical model of hydrogen-induced intergranular fracture in steel. J Mech Phys Solids. 2010;58(2):206-226. https://doi.org/10.1016/j.jmps.2009.10.005
48. Yamabe J, Matsumoto T, Matsuoka S, Murakami Y. A new mechanism in hydrogen-enhanced fatigue crack growth behavior of a 1900-MPa-class high-strength steel. Int J Fract. 2012;177(2):141-162. https://doi.org/10.1007/s10704-012-9760-9
49. Murakami Y, Takagi T, Wada K, Matsunaga H. Essential structure of S-N curve: prediction of fatigue life and fatigue limit of defective materials and nature of scatter. Int J Fatigue. 2021;146:106138. https://doi.org/10.1016/j.ijfatigue.2020.106138
50. Furuya Y, Matsuoka S, Abe T, Yamaguchi K. Gigacycle fatigue properties for high-strength low-alloy steel at 100 Hz, 600 Hz, and 20 kHz. Scr Mater. 2002;46(2):157-162. https://doi.org/10.1016/s1359-6462(01)01213-1
51. Görzen D, Schwich H, Blinn B, et al. Influence of Cu precipitates and C content on the defect tolerance of steels. Int J Fatigue. 2021;144:106042. https://doi.org/10.1016/j.ijfatigue.2020.106042
52. Spriestersbach D, Grad P, Kerscher E. Threshold values for very high cycle fatigue failure of high-strength steels. Fatigue Fract Eng Mater. 2017;40(11):1708-1717. https://doi.org/10.1111/fle.12682
53. Murakami Y, Nomoto T, Ueda T. Factors influencing the mechanism of superlong fatigue failure in steels. Fatigue Fract Eng Mater Struct. 1999;22(7):581-590. https://doi.org/10.1046/j.1460-2695.1999.00187.x
54. Sande M, Müller T, Lebahn J. Influence of mean stress and variable amplitude loading on the fatigue behaviour of a high-strength steel in VHCF regime. Int J Fatigue. 2014;62:10-20. https://doi.org/10.1016/j.ijfatigue.2013.04.015
55. Kovacs S, Beck T, Singheiser L. Influence of mean stresses on fatigue life and damage of a turbine blade steel in the VHCF-regime. Int J Fatigue. 2013;49:90-99. https://doi.org/10.1016/j.
ijfatigue.2012.12.012
56. Schönauer BM, Yanase K, Endo M. VHCF properties and fatigue limit prediction of precipitation hardened 17-4PH stainless steel. Int J Fatigue. 2016;88:205-216. https://doi.org/10.1016/j.
ijfatigue.2016.03.034
57. Spriestersbach D, Brodyanski A, Lösch J, Kopnarsi M, Kerscher E. Very high cycle fatigue of high-strength steels: crack initiation by FGA formation investigated at artificial defects. Proc Struct Integr. 2016;2:1101-1108. https://doi.org/10.
1016/j.prostr.2016.06.141
58. Stanzl-Tschegg S, Schönauer B. Near-threshold fatigue crack propagation and internal cracks in steel. Proc Eng. 2010;2(1):1547-1555. https://doi.org/10.1016/j. proceng.2010.03.167
59. Ooka T, Yoshimoto K, Nakamura T. Effects of vacuum environment on fatigue properties of high strength steel with a small defect. Trans Ipn Soc Mech Eng A. 2013;79(806):1545-1549. https://doi.org/10.1299/kikaia.79.1545
60. Ogawa T, Stanzl-Tschegg SE, Schönauer BM. A fracture mechanics approach to interior fatigue crack growth in the very high cycle regime. Eng Fract Mech. 2014;115:241-254. https://doi.org/10.1016/j.engfracmech.2013.11.007
61. Yoshinaka F, Nakamura T, Nakayama S, Shiozawa D, Nakai Y, Uesugi K. Non-destructive observation of internal fatigue crack growth in Ti–6Al–4V by using synchrotron radiation μCT imaging. Int J Fatigue. 2016;93:397-405. https://doi.org/10.
1016/j.ijfatigue.2016.05.028
62. Hong Y, Lei Z, Sun C, Zhao A. Propensities of crack interior initiation and early growth for very-high-cycle fatigue of high strength steels. Int J Fatigue. 2014;58:144-151. https://doi.org/10.
1016/j.ijfatigue.2013.02.023
63. Shiozawa K, Lu L, Ishihara S. S–N curve characteristics and subsurface crack initiation behaviour in ultra-long life fatigue
of a high carbon-chromium bearing steel. *Fatigue Fract Eng Mater Struct*. 2001;24(12):781-790. https://doi.org/10.1046/j.1460-2695.2001.00459.x

64. Tanaka K, Akiniwa Y. Fatigue crack propagation behaviour derived from S–N data in very high cycle regime. *Fatigue Fract Eng Mater Struct*. 2002;25(8–9):775-784. https://doi.org/10.1046/j.1460-2695.2002.00547.x

65. Pippan R, Tabernig B, Gach E, Riemelmoser F. Non-propagation conditions for fatigue cracks and fatigue in the very high-cycle regime. *Fatigue Fract Eng Mater Struct*. 2002;25(8–9):805-811. https://doi.org/10.1046/j.1460-2695.2002.00568.x

66. Yoshinaka F, Nakamura T. Effect of vacuum environment on fatigue fracture surfaces of high strength steel. *Mech Eng Lett*. 2016;2:15-00730. https://doi.org/10.1299/mel.15-00730

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