



Near-threshold fatigue crack growth and fatigue strength of ER70S-6 steel processed by wire arc additive manufacturing

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Abstract

In this work, fatigue crack growth (FCG) tests were conducted to determine the threshold stress intensity factor (ΔK_{th}) of ER70S-6 steel processed by wire arc additive manufacturing (WAAM). Samples were orientated with the crack direction (CD) both parallel ($//$) and normal ($\perp$) to build direction (BD). In addition, tests were conducted to obtain data of S–N fatigue strength and cycles to failure (N_f). Following S–N testing, the sizes of defects were determined. The resulting S–N data, together with the measured defect sizes and ΔK_{th} values, were subsequently analysed using the Kitagawa–Takahashi (K–T) diagram approach. Applying $R=0.1$, the ΔK_{th} value for CD//BD samples at $\sim 4.5 \text{ MPa } \sqrt{\text{m}}$ has been found to be slightly lower than the ΔK_{th} value for CD $\perp$ BD samples at $\sim 5.0 \text{ MPa } \sqrt{\text{m}}$. Some elongated α -ferrite grains orientated close to BD have been identified, suggesting that the cementite containing pearlite colonies may be locally aligned with the BD, reducing fatigue crack resistance along BD and thereby lowering ΔK_{th} . For both CD//BD and CD $\perp$ BD samples, the effect of the defect size up to $520 \text{ } \mu\text{m}$ affecting N_f can be explained using the K–T diagram approach. It is also shown that, when the effect of defects is excluded, the endurance limit of CD $\perp$ BD samples is slightly higher than that of CD//BD samples, consistent with ΔK_{th} (CD $\perp$ BD samples) $>$ ΔK_{th} (CD//BD samples).

Keywords Additive manufacturing defects · Damage tolerance · Build orientation effect

1 Introduction

Metal additive manufacturing (MAM) processes have advanced, matured and now increasingly adopted by a range of industries [1, 2]. Powder bed fusion (PBF) is the most widely adopted of these processes and enables the

fabrication of components with high geometric complexity. Nevertheless, part size is typically limited, and both build and production rates are relatively low [2]. On the other hand, although its geometric complexity is more limited than that achievable with powder bed fusion (PBF), wire arc additive manufacturing (WAAM) offers the capability to produce large-scale components at high deposition rates resulting in enhanced productivity. Owing to this advantage and its potential for large-scale industrial MAM, various aspects of WAAM have been extensively studied over the last decade and comprehensively reviewed in numerous research outputs in recent years [3–17]. During this period, the successful fabrication of WAAM parts, from small to very large in size has been widely demonstrated, as documented in these review articles [3, 6–8, 12–14].

The most widely adopted WAAM process is based on the advanced gas metal arc welding (GMAW) technique, particularly the cold metal transfer (CMT) mode, which provides precise control of heat input and material transfer [11]. Although WAAM can be employed to fabricate parts from a range of metallic alloys, provided they possess adequate weldability, steels remain the most extensively used

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alloys owing to their applicative diversity. However, fatigue performance is a critical consideration in many engineering applications and civil infrastructure designs, such as bridges. Consequently, considerable research attention has been directed towards the fatigue behaviour of WAAM-fabricated steels in recent years, particularly those produced using the widely employed ER70S-6 steel wire [18–27]. In these studies conducted at a stress ratio of $R=0.1$, the reported fatigue limits, expressed in terms of stress amplitude (σ_A), vary considerably, ranging from 115 MPa with run-out criterion of 7×10^6 cycles [18] to 180 MPa with run-out criterion of 5×10^6 cycles [22]. It should also be noted that the reported ultimate tensile strength (UTS) values exhibit substantial variation, ranging from 470 MPa [18] to 587 MPa [22]. Furthermore, an orientation dependence of fatigue performance has been observed. Specifically, fatigue limits of $\sigma_A = 173$ MPa and $\sigma_A = 193$ MPa have been reported for loading directions perpendicular ($\perp$) and parallel ($\parallel$) to the build direction (BD), respectively, with both values corresponding to a run-out criterion of 2×10^6 cycles [20].

Defects inherent to the WAAM process have been identified on the fracture surfaces of fatigue-tested WAAM ER70S-6 steel specimens and have been shown to act as crack initiation sites [18, 20, 23]. However, the quantitative influence of defect characteristics, particularly their size and location within the specimen, on the fatigue life (N_f) has not been systematically evaluated. Recent reviews [28–33]

have comprehensively examined the effects of process-induced defects on the fatigue strength of PBF processed alloys. These studies have demonstrated the effectiveness of the Kitagawa–Takahashi (K–T) diagram as a framework for quantitatively assessing the influence of defect size on fatigue behaviour. This approach requires the knowledge of the fatigue crack propagation threshold, expressed as the threshold stress intensity factor (ΔK_{th}). Importantly, ΔK_{th} defines the condition below which fatigue cracks associated with defects are unable to propagate. A number of fatigue crack growth (FCG) studies [22, 25, 26] on WAAM ER70S-6 steel in the as-built state have reported fatigue crack growth behaviour in terms of the Paris–Erdogan relationship, which describes the crack growth rate (da/dN) as a function of the stress intensity factor range (ΔK). Notably, Mendez-Morales et al. [26] reported that both the FCG growth mechanism and growth rate are independent of material condition (i.e. as-built or heat treated). However, a common limitation across these studies is the absence of an experimentally determined threshold stress intensity factor ΔK_{th} .

In this work, FCG tests were conducted on samples taken from as-built WAAM ER70S-6 steel plates to determine the threshold stress intensity factor (ΔK_{th}). Test samples were orientated in both the CD//BD (crack direction parallel to build direction) and CD $\perp$ BD configurations. The corresponding crack paths and microstructures were subsequently examined to assess whether orientation-dependant microstructural features contribute to variations in ΔK_{th} . Following the determination of ΔK_{th} , fatigue life (S–N testing) was performed to provide insight into the defect tolerance of as-built WAAM ER70S-6 steel. Fracture surfaces from the S–N tests were analysed to quantify the size of defects associated with failure. Finally, the Kitagawa–Takahashi (K–T) diagram approach was employed to evaluate the influence of WAAM-induced defects on the fatigue strength of the material.

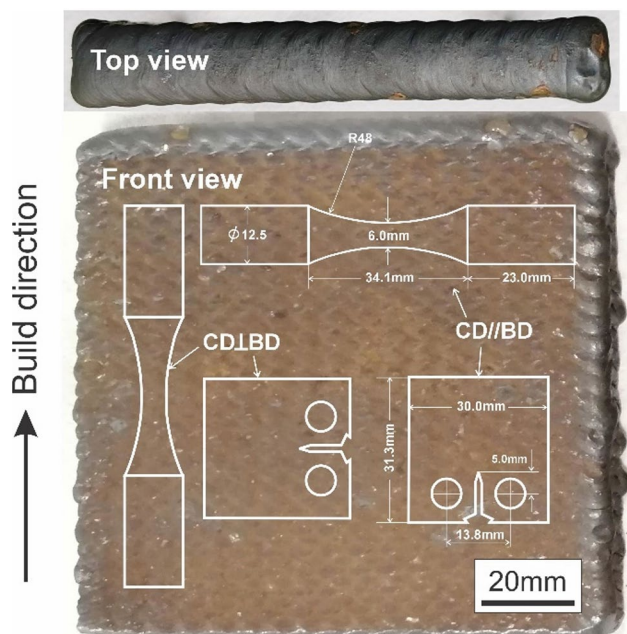


Fig. 1 Illustration of FCG and S–N samples, including their geometries and dimensions (CT samples thickness: 6 mm), extracted from a WAAM-processed ER70S-6 plate

2 Experimental procedures

2.1 Steel and WAAM conditions

WAAM plates were built using a Fronius Cold Metal Transfer (CMT) welding power source (TransPulsSynergic 4000) attached to a robotic arm (ABB IRB2400L robot). ER70S-6 steel (containing nominally 0.07% C, 0.85% Si and 1.45% Mn) wire with a diameter 1.2 mm was used. Plates with dimensions approximately $100 \times 100 \times 15 \text{ mm}^3$ (Fig. 1) were built on a mild steel substrate. Welding toolpath is a square

wave pattern with 14.5 mm width and 3 mm horizontal pitch. A welding speed of 20 mm/s and wire feed rate of 7 m/min were used. A resulting layer height of 2.1 mm was achieved with welding current and voltage targeted at 214A and 16.2 V, respectively, and correspondently adjusted by the control program set by the CMT machine manufacturer [34]. Shielding gas was a mixture of 2% Oxygen, 12% Carbon Dioxide, and balance of Argon. Interpass temperature between layers was monitored and maintained below 200 °C with a variable cooling time, nominally around 200 s.

2.2 Fatigue crack growth and S–N fatigue testing

Compact-test (CT) samples were wire-cut from the WAAM plates with test sample size following ASTM standards E647 [35]. As illustrated in Fig. 1, samples were taken in two orientations, CD//BD and CD⊥BD. The sample thickness is 6 mm. Samples were tested using an MTS Landmark Servo hydraulic system, with load frequency set at 20 Hz and $R=0.1$. Crack length (a) was monitored using the compliance method through the MTS Multipurpose Elite software in conjunction with an Epsilon clip-on gauge, utilising the equation:

$$\frac{a}{W} = 1.000 - 4.500u + 13.157u^2 - 172.551u^3 + 879.944u^4 - 1514.671u^5 \quad (1a)$$

where W is the width (from the centre line of the loading pins to the back edge of the FCG sample), u is the compliance parameter:

$$u = \frac{1}{\sqrt{E \cdot B \cdot \frac{\Delta\delta}{\Delta F} + 1}} \quad (1b)$$

where E is the young's modulus (210 GPa), B is the thickness of the sample (6 mm), $\Delta\delta$ is the range of the crack opening displacement and ΔF is the applied load range. During testing, the ΔK value was calculated automatically by the MTS software using the equation also specified in ASTM E647, which relates ΔK to the applied load, sample thickness, sample width, and crack length. Following the standard, a 1 mm pre-crack was initiated before the FCG test. K-decreasing started at 9–10 MPa m^{1/2}. A test was stopped when da/dN was approaching 10^{−10} m/cycle. Two K-decreasing tests for each sample orientation, thus four in total, were conducted to determine the ΔK_{th} values. For

one sample, K-increasing was also carried out following the K-decreasing test.

S–N fatigue life tests were conducted using the same MTS fatigue testing system to generate specimens for evaluating the fatigue tolerance of WAAM-induced defects. The testing procedure was performed in accordance with ASTM E466 [36]. Samples with both CD//BD and CD⊥BD orientations were cut from the WAAM plates, as illustrated in Fig. 1. Specimen geometry, including the radius of curvature, gauge diameter, and reduced section length, was designed in accordance with the ASTM standard to minimise stress concentration effects. Consistent with the standard, the machined samples were subjected to cylindrical grinding followed by longitudinal polishing as the final surface preparation step. Fatigue tests were conducted at a loading frequency of 20 Hz and a stress ratio of $R=0.1$. The applied maximum stress was from 540 to 450 MPa.

2.3 Post testing examination and analysis

After fatigue tests, metallographic and fractographic examinations were conducted on the samples using a Hitachi SU-70 scanning electron microscope (SEM) equipped with electron backscatter diffraction (EBSD) capability (Thermo Scientific NSS—Noran System 7). EBSD orientation mapping was performed to visualise whether crack propagation was transgranular or intergranular. For EBSD scanning, an accelerating voltage of 20 kV was used. EBSD indexing was based on the generic/standard BCC crystal structure provided in the NSS software, as α -ferrite is a BCC phase. Selected FCG tested samples were sectioned with the sectioned face including the crack path, mounted, and polished following the conventional metallography procedure. For EBSD preparation, electrolytic polishing was conducted in a 10% perchloric acid solution diluted by ethanol for 20 s at 25 V. For normal metallographic observation, the samples were etched using 2% Nital. Vickers hardness measurement was made on polished sample surfaces. For fatigue tolerance estimation, defect areas were measured using ImageJ on images taken during fractographic examination of the S–N fatigue tested samples.

3 Results and discussion

3.1 Fatigue crack growth and ΔK_{th}

Although the present work focuses on K-decreasing tests to determine ΔK_{th} values, one K-increasing test was conducted following the K-decreasing test, using the same

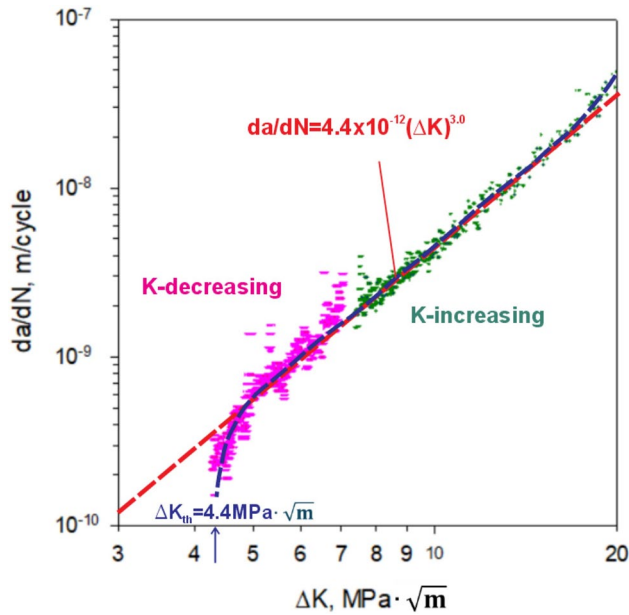


Fig. 2 Fatigue crack growth rate (da/dN) as a function of the stress intensity factor (ΔK) for a CD//BD sample. The dashed red line represents the Paris regime, while the dashed blue curve is used to determine the threshold stress intensity factor range (ΔK_{th}) at $da/dN=10^{-10}$ m/cycle

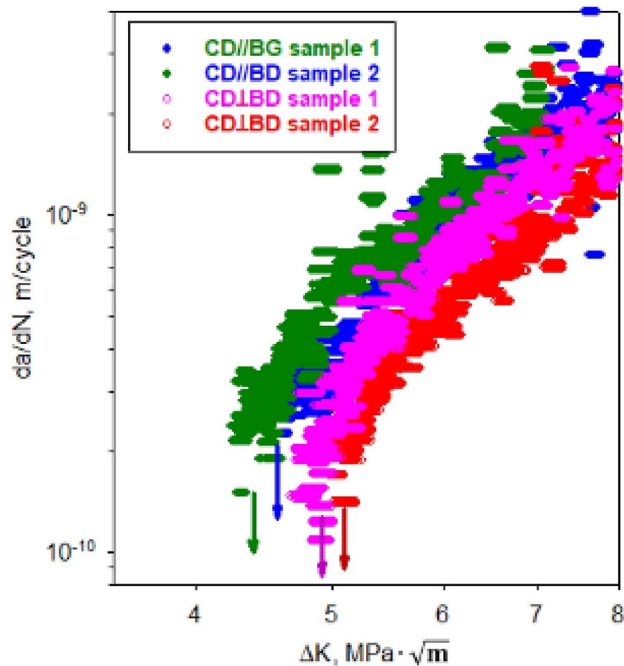


Fig. 3 Fatigue crack growth rate (da/dN) as a function of the stress intensity factor (ΔK) from the decreasing K tests for CD//BD ($\Delta K_{th}=4.4$ and 4.6 MPa $m^{1/2}$), and CD⊥BD samples ($\Delta K_{th}=4.9$ and 5.1 MPa $m^{1/2}$)

sample. Fatigue crack growth data for both K-decreasing and K-increasing tests of the sample is presented in Fig. 2. A straight (red and broken) line is drawn to provide a visual representation of the K-decreasing results for $\Delta K > 5$ MPa $m^{1/2}$, as well as a substantial portion of the K-increasing data. This line corresponds to the Paris regime and corresponds to Paris–Erdogan parameters of $C=4.4 \times 10^{-12}$ and $m=3.0$ for the Paris equation using da/dN in m/cycle and ΔK in MPa $m^{1/2}$. These values are consistent with values previously reported for WAAM ER70S-6 steel and summarised by Mendez-Morales [26]: C ranges from 1.6×10^{-11} m/cycle to 1.0×10^{-13} m/cycle and m ranges from 2.6 to 3.4 for $R=0.1$. The second curve (blue and broken) in Fig. 2 is the visual representation tracing the experimental data in the low ΔK regime from the ΔK -decreasing test. In accordance with ASTM E647, the threshold stress intensity factor range (ΔK_{th}) was determined by extrapolating the traced curve to a crack growth rate of $da/dN=10^{-10}$ m/cycle. As illustrated Fig. 2, the resulting ΔK_{th} for this sample is 4.4 MPa $m^{1/2}$.

As the primary focus of this study is the determination of ΔK_{th} and its implications for the fatigue tolerance of WAAM ER70S-6 steel, particular attention is given to fatigue crack growth data in the regime approaching $da/dN=10^{-10}$ m/cycle. Accordingly, Fig. 3 presents the fatigue crack growth data obtained from four ΔK -decreasing tests, comprising of two samples in the CD//BD orientation and two in the CD⊥BD orientation. The apparent scatter in the data arises from the high data acquisition rate and the absence of data smoothing. Confident determination of ΔK_{th} values can be obtained from the data in Fig. 3, as the measured crack growth rates extend to values approaching $da/dN=10^{-10}$ m/cycle. As indicated by the arrows in Fig. 3, the measured ΔK_{th} values are 4.4 and 4.6 MPa $m^{1/2}$ for the two CD//BD samples and 4.9 and 5.1 MPa $m^{1/2}$ for the CD⊥BD samples. It should be noted that, although two values have been reported for each condition, the number of tests is limited. Although FCG behaviour in Region II is generally considered insensitive to microstructural changes, microstructural features can influence FCG behaviour in Region I and the ΔK_{th} [37]. Therefore, in the following, microstructural characterisation of both the as-built WAAM material and the FCG-tested samples was undertaken to investigate whether orientation-dependent microstructural features contribute to the observed differences in FCG behaviour and ΔK_{th} between the two build orientations.

3.2 Microstructure and crack growth paths

The typical grain structure of the samples is shown in Fig. 4. At higher magnification (Fig. 4 right), the

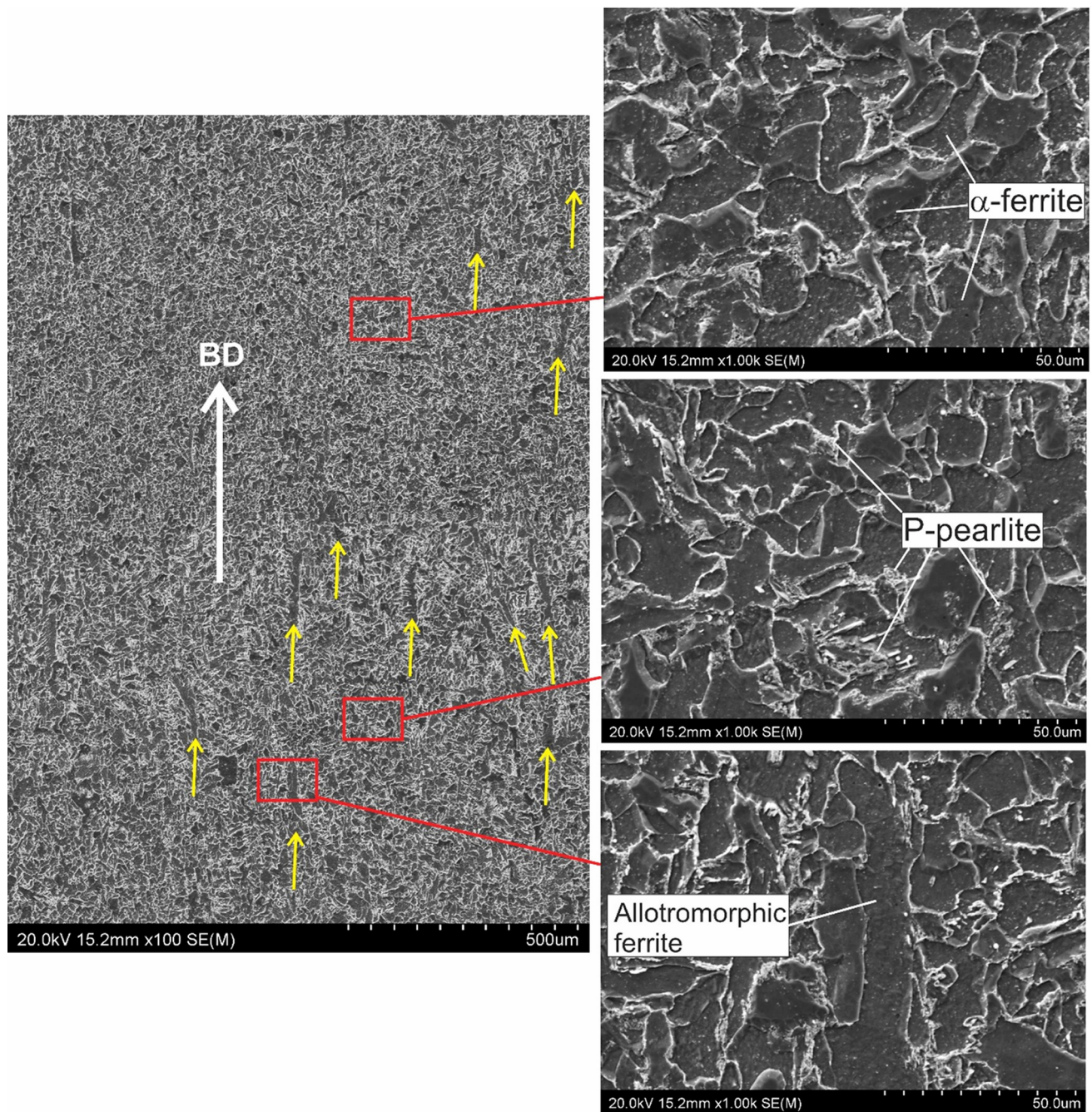


Fig. 4 SEM micrographs of WAAM ER70S-6 steel showing weak columnar growth at low magnification and predominantly equiaxed α -ferrite grains with grain-boundary pearlite at high magnification. Arrows indicate elongated α -ferrite grains

microstructure is dominated by α -ferrite grains, with P-pearlite (ferrite+cementite) forming intergranularly. This microstructure is consistent with that widely reported for WAAM-fabricated ER70S-6 steel [38–41]. During WAAM solidification of ER70S-6 steel, δ -ferrite columnar grains are expected to grow preferentially along the build direction (BD), followed by transformation to γ -austenite and

subsequently to α -ferrite and P-pearlite upon cooling. The relatively low carbon content of the alloy results in a limited volume fraction of pearlite. As shown in the top-right and mid-right micrographs of Fig. 4, the ferrite grains are predominantly equiaxed, although some elongated α -ferrite grains are also observed (an example is shown in Fig. 4 lower-right). These elongated grains are generally aligned

close to the build direction (as seen in Fig. 4 left). Several such features are highlighted by yellow arrows in the low-magnification micrograph.

The elongated grains are identified as allotriomorphic α -ferrite [40]. These grains form along prior γ -austenite grain boundaries during $\gamma \rightarrow \alpha$ transformation, as widely reported in steel welding literature [42, 43]. However, the details of this transformation during the WAAM process have received limited attention in the literature. During solidification, the columnar δ -ferrite grains are expected to transform into columnar γ -austenite grains, followed by the formation of allotriomorphic α -ferrite along the columnar γ -austenite grain boundaries. Consequently, the newly formed allotriomorphic α -ferrite grains are predicted to be preferentially aligned with the build direction (BD). Subsequent thermal cycling associated with the deposition of the next ($n+1$) layer is expected to induce the $\alpha \rightarrow \gamma \rightarrow \alpha$ phase transformation in the HAZ of the underlying n -th layer. This repeated transformation may largely eliminate the elongated allotriomorphic α -ferrite grains, leaving only a limited number of residual elongated α -ferrite grains in the final microstructure.

As shown in Fig. 4 (left), elongated α -ferrite grains are more prevalent in the lower portion of the micrograph than in the upper portion. This observation is consistent with the thermal conditions experienced during deposition of the subsequent $n+1$ layer. Specifically, regions within the heat-affected zone (HAZ) of the underlying n -th layer that are located further from the overlying melt pool are subjected to lower peak temperatures. As a result, the $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation may be incomplete in some of the original allotriomorphic α -ferrite grains, leading to the retention of a greater number of elongated α -ferrite grains in these regions. Variations in grain morphology across individual WAAM layers have been commonly reported in low-magnification micrographs of WAAM ER70S-6 steel [41, 44, 45]. Similarly, the high-magnification SEM micrographs presented by Laghi et al. [41] or the so-called “layer core” and “interlayer” regions exhibit features comparable to those observed in Fig. 4. In both studies, one sublayer region is characterised by predominantly equiaxed α -ferrite grains, whereas the other contains locally elongated α -ferrite grains, reflecting the heterogeneous thermal history associated with the WAAM process.

Vickers' hardness values measured on four polished samples, two with the cross-sectional plane parallel to build direction (P//BD) and two with the cross-sectional plane normal to build direction (P $\perp$ BD), are presented in Table 1. For each sample, 22 measurements were made, thus totaling 44 for either P//BD or P $\perp$ BD condition. Although the standard deviation values show the hardness values to have varied 4–5%, the mean values together with the low standard error values suggest that hardness in P//BD condition (Hv-172.5) is slightly higher than that of the P $\perp$ BD condition (Hv167.2).

EBSD pattern quality images and orientation maps containing the fatigue crack paths are presented in Fig. 5. Consistent with the SEM observations, the microstructure is characterised predominantly by equiaxed grains, with most grains measuring approximately 10–15 μm in size. An elongated α -ferrite grain is also evident, as indicated in Fig. 5b. Examination of the crack paths in both the CD//BD and CD $\perp$ BD samples reveals that crack propagation occurred through a combination of transgranular and intergranular mechanisms. Regions of intergranular crack growth are highlighted by white arrows in the figures. Since crack propagation can locally alternate between these two modes, pronounced crack branching is not observed in either orientation. Nevertheless, cementite particles located along grain boundaries are known to be relatively brittle and may facilitate intergranular crack propagation. Furthermore, the presence of elongated α -ferrite grains aligned approximately with the build direction (BD) may result in a more continuous distribution of grain-boundary cementite particles along this direction.

It is noted in Fig. 5b that there is a clear crack path deflection locally near the crack tip in the CD $\perp$ BD sample. Before that deflection, the crack locally grew in an intergranular manner as pointed to by the two arrows in the higher magnification EBSD orientation map. The crack path deflected at a high angle and crack growth was then in a transgranular manner through the two grains to form that deflecting step. Into the third grain the transgranular crack growth in the same direction as before must become difficult and therefore deflected back to grow in the direction more normal to the tensile loading direction. Thus, a higher fraction of transgranular crack path may result in a slightly rougher crack path. As has been indicated by the white arrows pointing to the intergranular crack growth locally in both Fig. 5a and b, a higher number of the arrows pointing to the intergranular crack growth can be identified in CD//BD condition than in CD $\perp$ BD direction. Thus, a roughness-induced crack close effect to increase crack growth resistance for the CD $\perp$ BD condition is possible. This, together with the microstructural arrangement, as explained in the last paragraph, may be the factors that have contributed to the slightly lower FCG

Table 1 Vickers hardness values of WAAM ER70S-6 measured on planes parallel to build direction (P//BD) and normal to build direction (P $\perp$ BD)

P//BD		P $\perp$ BD	
Mean	172.5	Mean	167.2
Standard deviation	9.1	Standard deviation	6.6
Standard error	1.4	Standard error	1.0

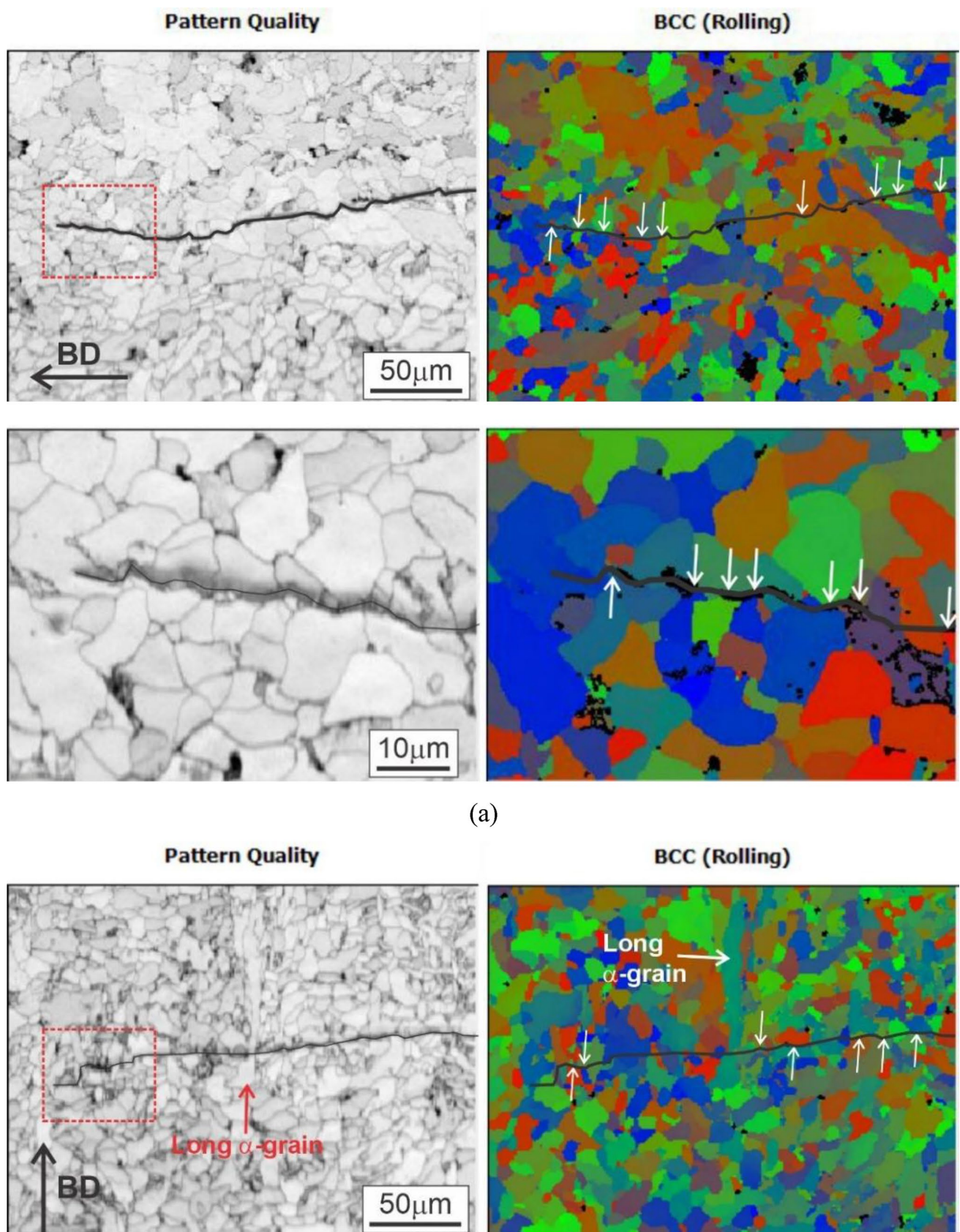


Fig. 5 EBSD maps near crack-tip regions (which are the ΔK_{th} regions—Region I), with the lower images showing higher-magnification views of the crack-tip areas outlined in the upper maps, for **a** CD//BD and **b**

CD⊥BD tested samples. White arrows indicate local crack propagation paths along grain boundaries

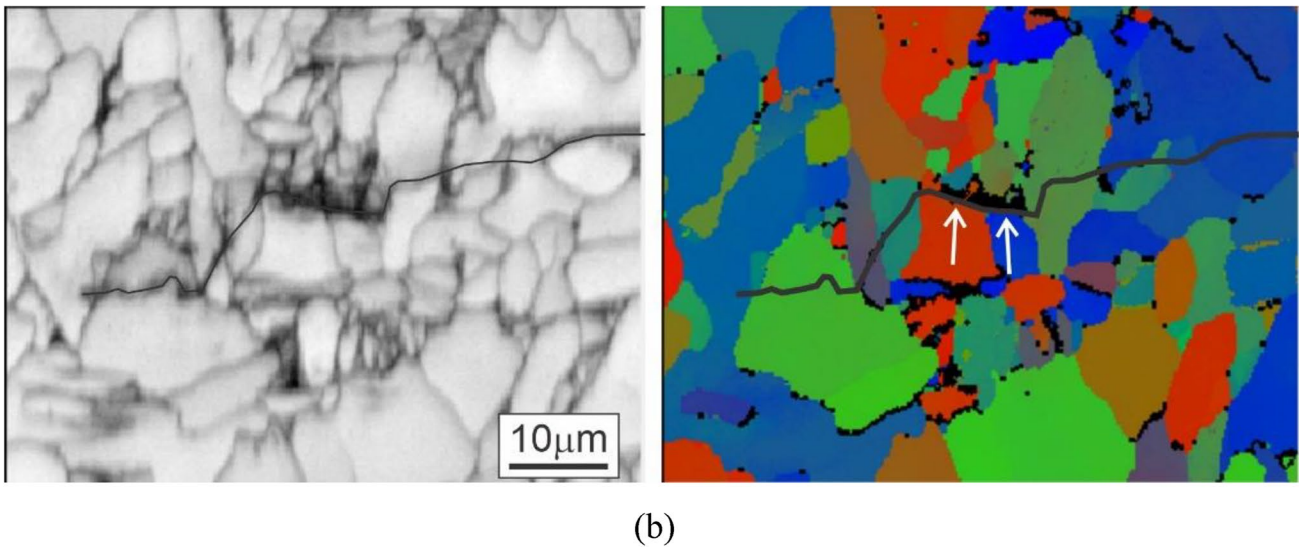
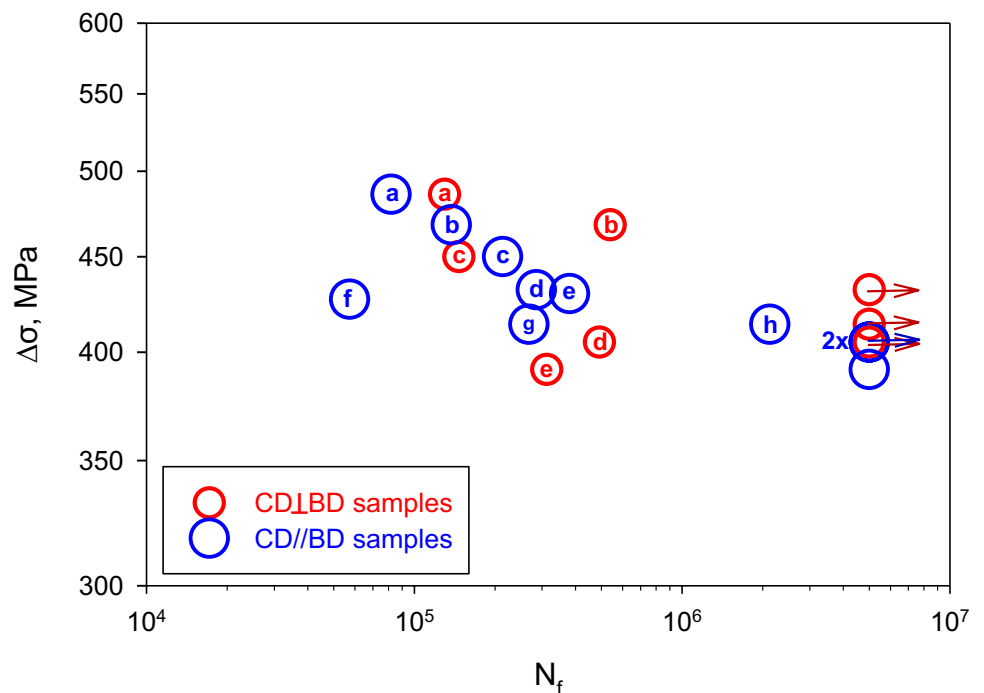


Fig. 5 (continued)

Fig. 6 S–N data for both CD//BD and CD⊥BD samples, with all non-run-out specimens labelled by letters corresponding to those used in the fracture surface examination analysis



resistance observed for the CD//BD samples, as reflected by their lower ΔK_{th} values ($\sim 4.5 \text{ MPa m}^{1/2}$) compared with those of the CD⊥BD samples ($\sim 5.0 \text{ MPa m}^{1/2}$).

3.3 Fatigue strength and fracture surfaces

As outlined in the Introduction, the objective of this study is to determine the ΔK_{th} of WAAM-processed ER70S-6 steel and evaluate the influence of process-induced defects on its fatigue behaviour. Thus, a number S–N fatigue tests have been conducted. The S–N fatigue life data, presented as stress range ($\Delta\sigma$) versus number of cycles to failure (N_f),

for the two sets of CD-BD build orientations are shown in Fig. 6. In this study, a specimen was classified as a run-out if failure had not occurred upon reaching $N = 5 \times 10^6$ cycles. As shown in Fig. 6, considerable scatter is evident in the S–N data. However, such variability is not uncommon in fatigue studies of additively manufactured materials [28–33]. Since fracture surface analyses were conducted and the ΔK_{th} were determined in the present work, the influence of WAAM-induced defects on fatigue life can be systematically assessed using the Kitagawa–Takahashi (K–T) diagram approach. Variations in defect size and location may therefore account for a significant portion of the observed

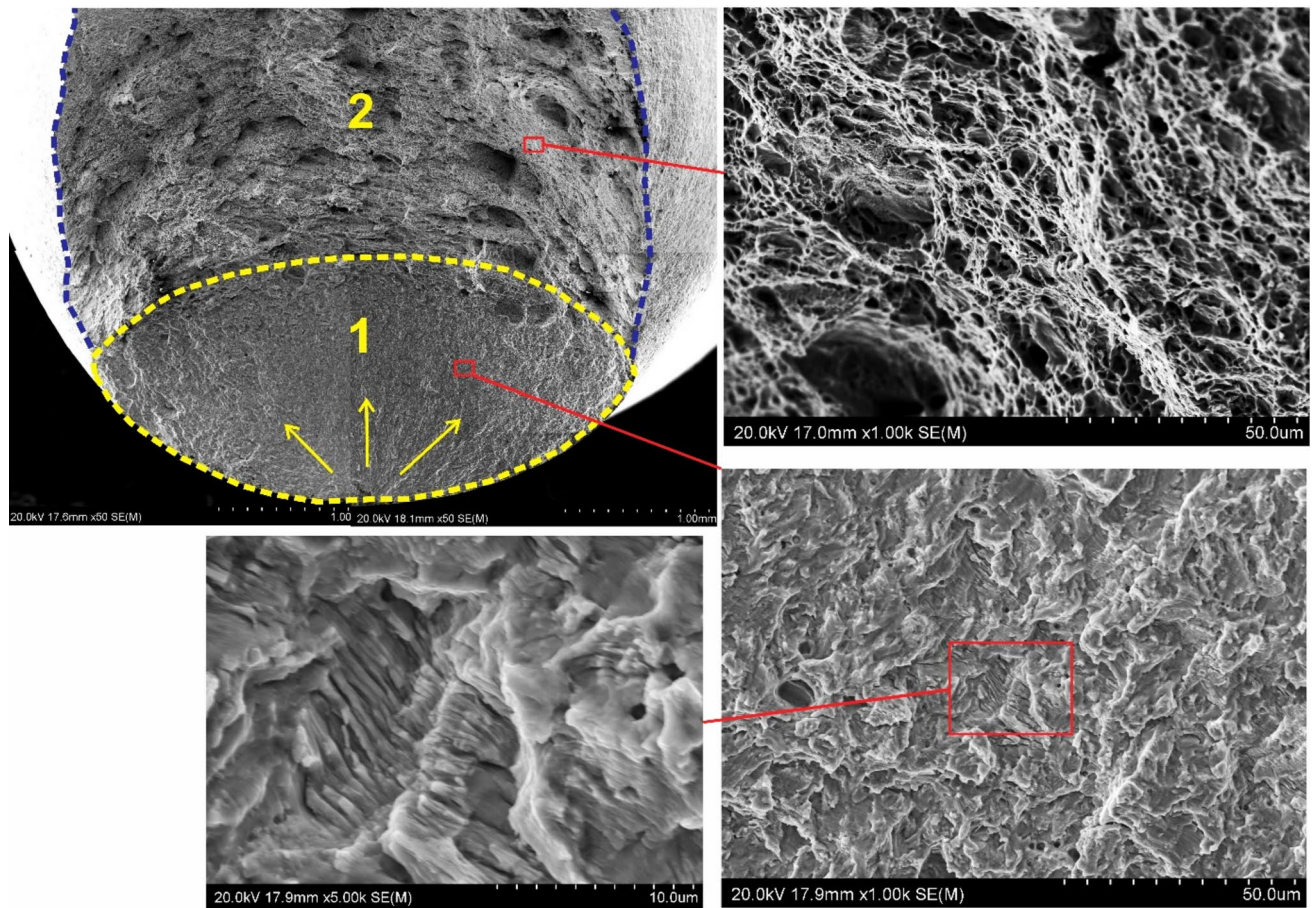


Fig. 7 Fractographs of the fracture surface of a selected sample, showing the identified crack initiation region (Area 1) with indicated crack growth directions and the fast fracture region (Area 2). In the high magnification (lower-left) fractograph, small fatigue striations can be seen

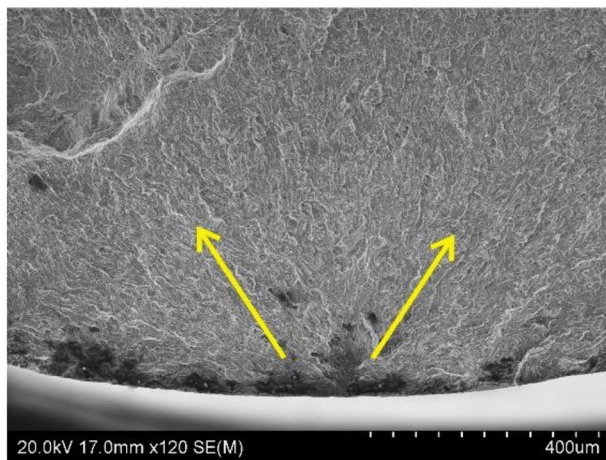
scatter in the S–N data. To facilitate this analysis, each failed specimen is identified by a letter in Fig. 6. These labelled specimens are subsequently plotted on the K–T diagram and individually evaluated with respect to their defect characteristics and fatigue performance.

Fractographs of a representative fracture surface are presented in Fig. 7, illustrating features common to all fractured specimens. As shown in the low-magnification fractograph on the left, the fracture surface can be divided into two distinct regions. Area 1 corresponds to the fatigue crack growth region and exhibits a relatively flat morphology characteristic of fatigue fracture. Area 2 represents the final rapid fracture region and is characterised by a much rougher surface appearance. Final fracture occurs when the fatigue crack in Area 1 reaches a critical size such that the remaining load-bearing cross-sectional area (Area 2) is unable to sustain the maximum applied load (F_{max}). Consequently, rapid fracture of the remaining ligament occurs. Higher-magnification examination of Area 1 (Fig. 7, lower left) reveals fatigue striations, confirming progressive fatigue crack propagation. Conversely, Area 2 (Fig. 7, upper right) exhibits a rough, dimpled morphology characteristic of ductile

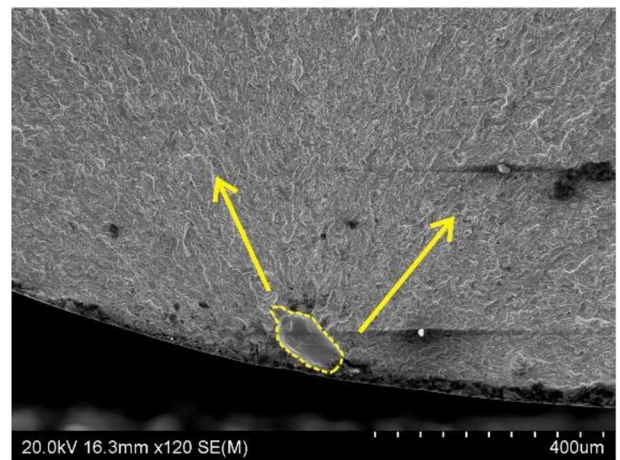
overload fracture. Notably, no pore defects were observed on the fracture surface, indicating that pore-induced crack initiation was not responsible for failure in this sample.

The fracture surfaces of all CD//BD and CD⊥BD samples are shown in Fig. 8 and Fig. 9, with particular attention given to the crack initiation and fatigue crack growth region (Area 1). A fractograph with a common magnification, $\times 120$ as stated in the bottom data bar with a 400 μm scale bar, has been used for each fracture sample in the figures in order to more readily compare one sample to another. However, in Fig. 8g, an additional fractograph with a higher magnification needs to be provided to show the very small pore in the crack initiation site. This is also the same in Fig. 9a. On the other hand, in Fig. 8h, the fractograph with the 400 μm scale bar does not show the edge of the sample. Thus, a lower magnification fractograph has thus been added. As has been explained in 2.3, defect areas were measured using ImageJ on fractographs. The square root of each measured defect area, $\sqrt{\text{area}}$, is stated in the captions of Figs. 8 and 9.

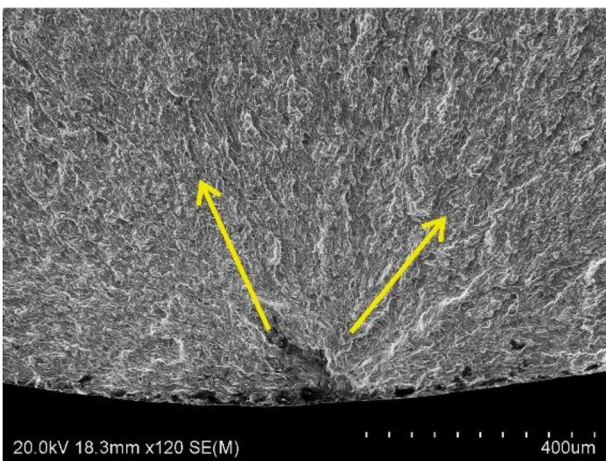
For sample-a of CD//BD condition (Fig. 8a), crack initiation was not associated with a pore defect, consistent with the fracture surface shown previously in Fig. 7. In sample-b



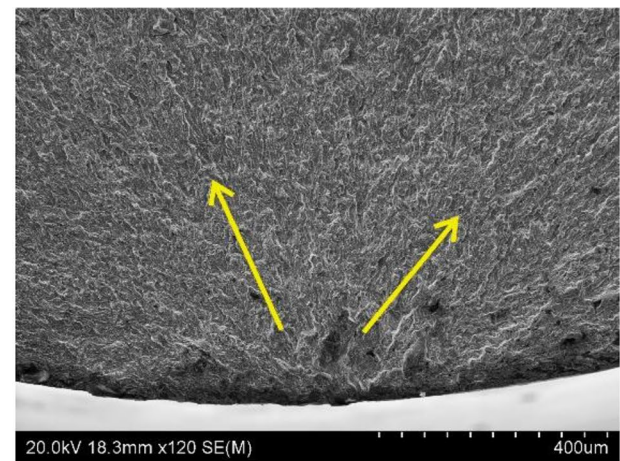
(a)



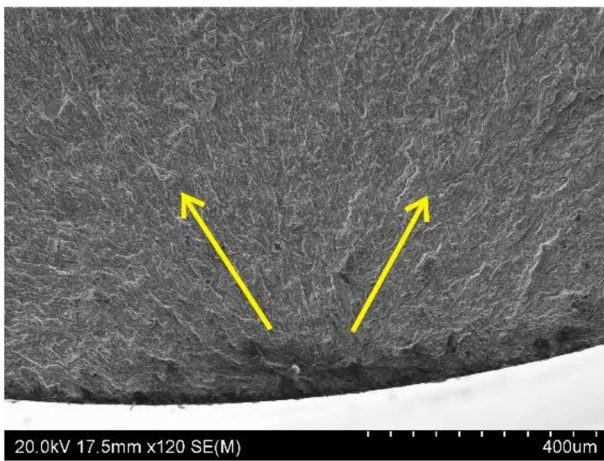
(b)



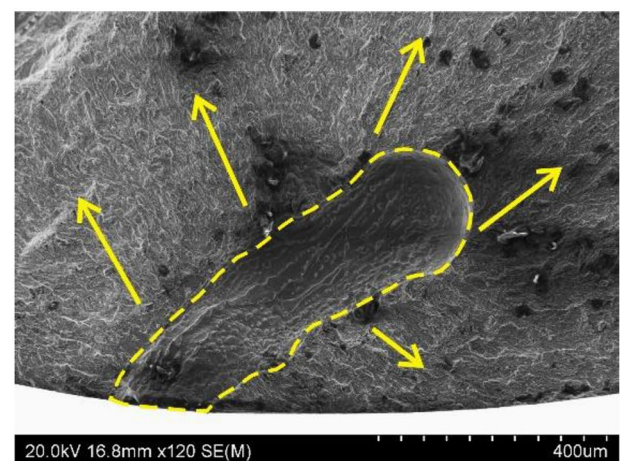
(c)



(d)



(e)



(f)

Fig. 8 Fracture surfaces of CD//BD samples (a–h), corresponding to the samples labelled in Fig. 6 with $\Delta\sigma$ and $\sqrt{area}$ =: **a** 486 MPa and 0 μm , **b** 468 MPa and 93 μm , **c** 450 MPa and 0 μm , **d** 432 MPa

and 0 μm , **e** 432 MPa and 0 μm , **f** 427 MPa and 349 μm , **g** 414 MPa and 44 μm , and **h** 414 MPa and 520 μm . Note: arrows indicate crack growth directions

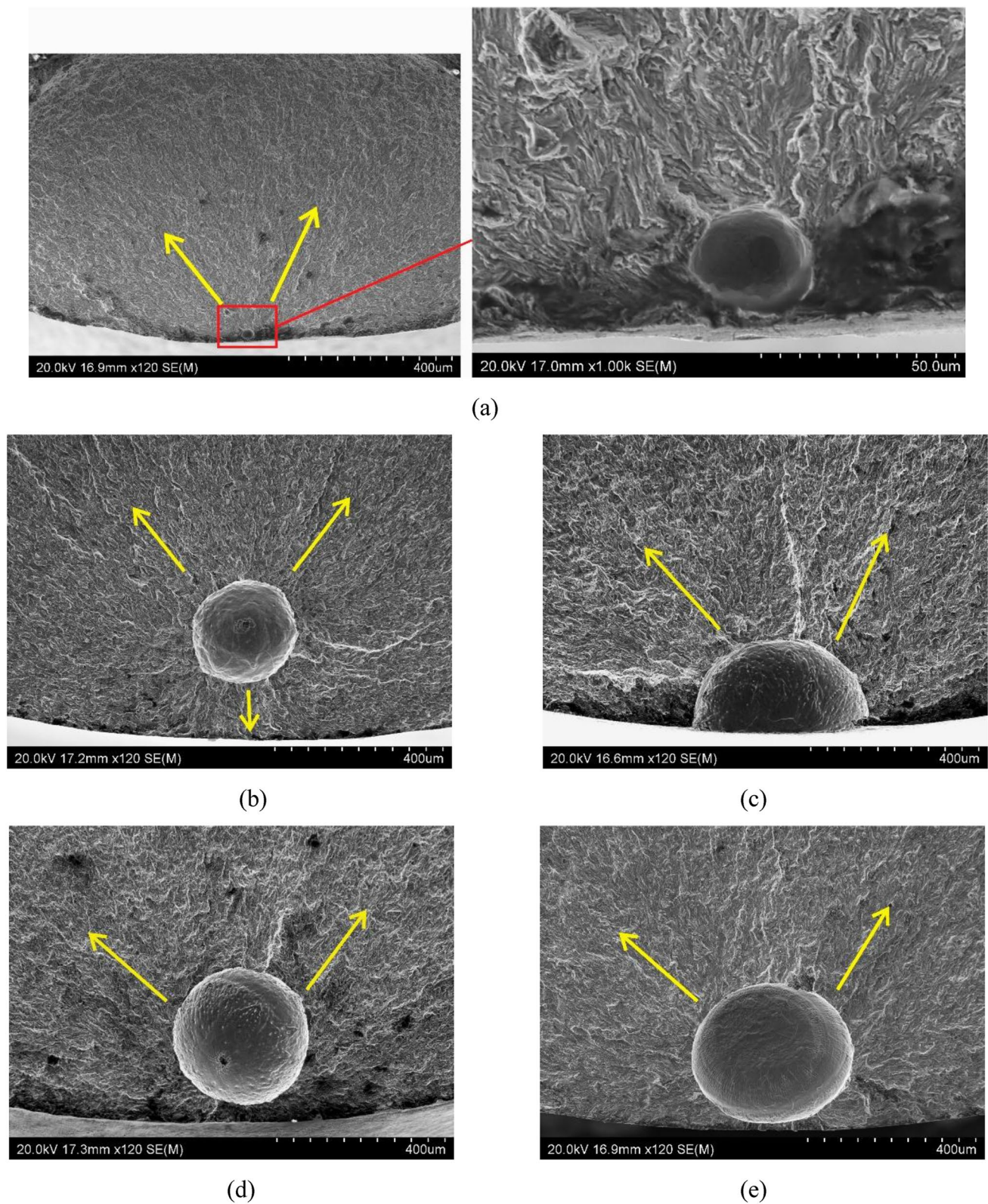


Fig. 9 Fracture surfaces of CDLBD samples (**a–d**), corresponding to the samples labelled in Fig. 6 with $\Delta\sigma$ = and $\sqrt{area}$ =: **a** 486 MPa and 24 μm , **b** 468 and 218 μm , **c** 450 and 223 μm , **d** 405 MPa and 288 μm , and **e** 391 MPa and 325 μm , Note: arrows indicate crack growth directions

(Fig. 8b), however, fatigue crack initiation occurred at a small pore defect. Similarly, no pore defects were apparent at the crack initiation sites of samples-c, sample-d, and sample-e. In contrast, sample-f contained a large, elongated pore defect that served as the fatigue crack initiation site. Despite being tested at an intermediate- $\Delta\sigma$, this specimen exhibited the shortest fatigue life (N_f) among all CD//BD samples (Fig. 6), indicating the significant influence of large defects on fatigue performance and contributing substantially to the scatter in the S-N data. For sample-g, only a very small defect was detected at the crack initiation site, whereas sample-h contained a readily identifiable large defect. Interestingly, although the defect in sample-h was considerably larger than that in sample-g, its fatigue life ($N_f=2.1 \times 10^6$) was substantially longer than that of sample g ($N_f=2.7 \times 10^5$) at the same stress range. This observation suggests that defect size alone is insufficient to explain the observed fatigue behaviour and that other factors, such as defect location and local stress intensity, must also be considered.

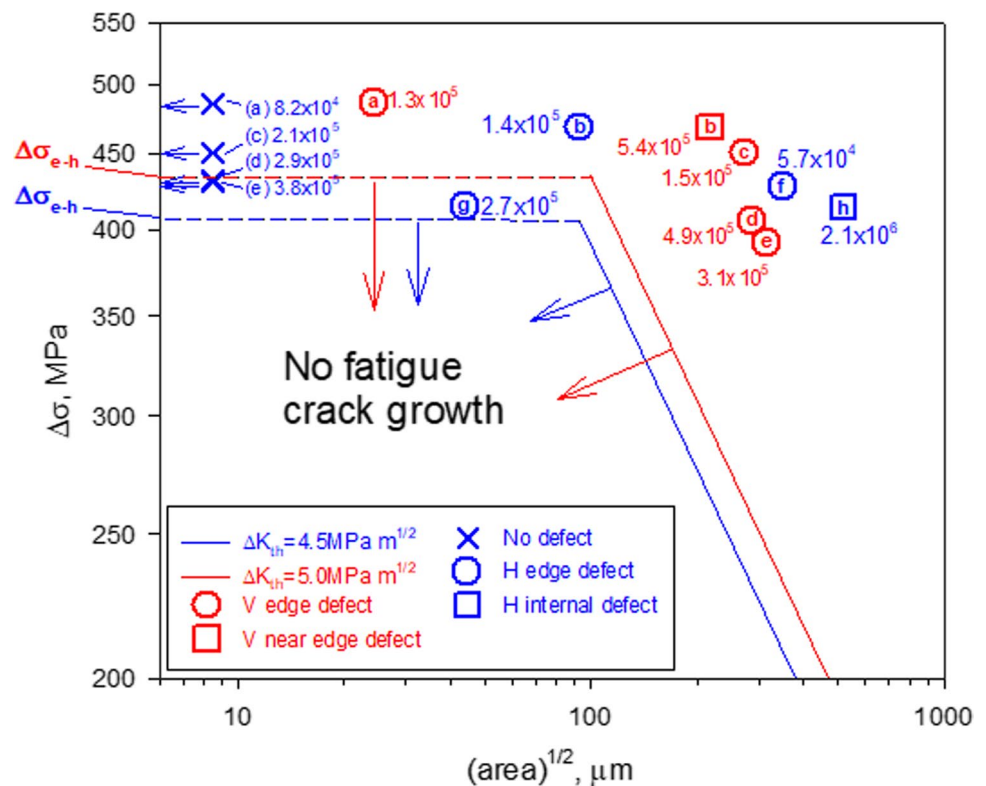
As shown in Fig. 9, pore defects were identified on the fracture surfaces of all non-run-out CD//BD built samples. In sample-a, sample-c, sample-d and sample-e, crack propagation from the pore defect into the surrounding fatigue crack growth region (Area 1) is clearly evident. These photographs demonstrate that fatigue cracks initiated at the pore defects and subsequently propagated to form the characteristic fatigue fracture surface. In sample-b (Fig. 9b), crack

growth is observed to radiate from the pore defect. The radiating crack pattern suggests that crack initiation occurred at the pore and propagated not only into the main fatigue crack growth region but also towards the lower specimen surface. Despite being tested at a relatively high stress range, this specimen exhibited a comparatively long fatigue life (N_f), contributing to the scatter observed in the S-N data. This behaviour suggests that, although the pore defect acted as the crack initiation site, its location beneath the surface may have reduced its detrimental effect on fatigue strength compared with an equivalent surface-connected defect. As has been explained in Introduction, pore defects are present in weld metals and are present in WAAM processed alloys. As ΔK_{th} values and pore defect size ($\sqrt{area}$) on fracture surface have been determined, the effect of pore defects on fatigue strength can now be estimated.

3.4 Effects of defects on fatigue strength

A Kitagawa-Takahashi (K-T) diagram can now be constructed after determining the ΔK_{th} , $\Delta\sigma$ and N_f and $\sqrt{area}$ values, as presented above, and used to evaluate how defects may affect the fatigue strength. The diagram was constructed using two threshold parameters [37]. The first is the fatigue (or endurance) limit, $\Delta\sigma_{e-h}$, which represents the stress range below which fatigue life is considered effectively infinite ($>5 \times 10^6$ cycles) in the absence of defect effects. As shown in Fig. 6, three specimens reached the run-out criterion.

Fig. 10 Kitagawa-Takahashi diagram based on the determined ΔK_{th} values, with $\Delta\sigma - \sqrt{area}$ test data also plotted



For the CD//BD orientated samples, the highest run-out stress range was 405 MPa, which was therefore taken as the defect-free endurance limit ($\Delta\sigma_{e-h}$). For the CD $\perp$ BD orientated samples, the highest run-out stress range ($\Delta\sigma_e$) was 435 MPa. Fractographic analysis (Fig. 9) revealed that samples-e and sample-d contained relatively large defects ($\sqrt{\text{area}}$ 315 and 288 μm respectively), indicating that their fatigue performance was significantly influenced by defect presence. Defect effects were also evident in samples-b and sample-c. Consequently, 435 MPa was adopted as the defect-free threshold ($\Delta\sigma_{e-h}$) for the CD $\perp$ BD samples. The resulting $\Delta\sigma_{e-h}$ values are represented in Fig. 10 by horizontal lines, with the blue line corresponding to the CD//BD orientation and the red line corresponding to the CD $\perp$ BD orientation.

The second threshold is based on the threshold stress intensity factor (ΔK_{th}), defines the influence of defect size on the reduction of the endurance limit ($\Delta\sigma_e$). This relationship is described using the Murakami equation, which relates fatigue strength to defect size and crack propagation threshold [37]:

$$\Delta\sigma_e = \frac{1}{Y} \cdot \frac{\Delta K_{th}}{\sqrt{\pi\sqrt{\text{area}}}} \quad (2a)$$

where $\Delta\sigma_e$ is the defect affected fatigue strength and Y is a geometry factor. For surface or near surface defects, $Y=0.65$ [37] and thus:

$$\Delta\sigma_e = 0.87 \cdot \Delta K_{th} \cdot (\sqrt{\text{area}})^{-1/2} \quad (2b)$$

In log scale:

$$\log \Delta\sigma_e = \log 0.87 \cdot \Delta K_{th} - \frac{1}{2} \log \sqrt{\text{area}} \quad (2c)$$

This relationship implies that, on a Kitagawa–Takahashi (K–T) diagram, the defect-affected $\Delta\sigma_e$ plotted against $\sqrt{\text{area}}$ on a logarithmic scale is represented by a straight line with a slope of -0.5 . For the CD//BD samples, in Fig. 3, the measured ΔK_{th} values are 4.4 and 4.6 MPa $\sqrt{\text{m}}$. Accordingly, a representative value of $\Delta K_{th}=4.5$ MPa $\sqrt{\text{m}}$ was adopted to construct the defect-sensitive branch of the K–T diagram. This line intersects the defect-free endurance limit ($\Delta\sigma_{e-h}=405$ MPa), yielding a critical defect size, $(\sqrt{\text{area}})_c$, of 94 μm (Fig. 10). Similarly, for the CD $\perp$ BD samples, the measured ΔK_{th} values are 4.9 and 5.1 MPa $\sqrt{\text{m}}$ (Fig. 3). Using a representative value of $\Delta K_{th}=5.0$ MPa $\sqrt{\text{m}}$, the defect-sensitive branch was constructed to intersect the defect-free endurance limit, $\Delta\sigma_{e-h}=435$ MPa. The resulting critical defect size was determined to be $(\sqrt{\text{area}})_c = 101\mu\text{m}$. The resulting K–T diagram is presented in Fig. 10. The

region below the threshold boundaries corresponds to the safe, no-fatigue-crack-growth (no-FCG) regime, where fatigue life is effectively unlimited. In contrast, specimens operating above these boundaries are susceptible to fatigue crack propagation and therefore exhibit finite fatigue lives under cyclic loading.

The $\Delta\sigma$ data (Fig. 6), paired with the corresponding measured $\sqrt{\text{area}}$ values (shown in Figs. 8 or 9), are plotted in the K–T diagram (Fig. 10), excluding the run-out samples, which are located within or near the boundary of the no-FCG region. The corresponding fatigue lives (N_f) for each data point are also indicated in Fig. 10. For the four CD//BD samples (a, c, d and e) where the arrows point to the left to indicate $\sqrt{\text{area}} \rightarrow 0$, the corresponding N_f values are also indicated in the figure. For the CD//BD sample-b (Fig. 8b), a small defect was observed to have initiated the fatigue crack. The measured $\sqrt{\text{area}}$ value is close to $(\sqrt{\text{area}})_c$; therefore, this defect would not be expected to significantly reduce the fatigue life or result in a low-to-medium N_f (at 1.4×10^5 cycles) unless $\Delta\sigma$ is relatively high. However, the applied $\Delta\sigma$ for this sample was high (468 MPa), exceeding the $\Delta\sigma_{e-h}$ value (405 MPa). Therefore, the observed low-to-medium fatigue life is considered reasonable. For the CD//BE sample-f, the large defect observed in Fig. 8 resulted in failure at an intermediate-to-low $\Delta\sigma$ level and produced the lowest N_f value among the tested samples (Fig. 6), leading to a significant data scatter. This effect is reflected in Fig. 10, where the $\Delta\sigma$ – $\sqrt{\text{area}}$ point for CD//BD sample-f is located a considerable distance from the boundary of the no-FCG region, demonstrating the pronounced influence of defect size on fatigue strength. For sample-g (Fig. 8g), $\sqrt{\text{area}} = 44\mu\text{m} < (\sqrt{\text{area}})_c$; therefore, the defect is not expected to significantly influence $\Delta\sigma_e$. However, for this sample, $\Delta\sigma = 414\text{MPa} > \Delta\sigma_{e-h} = 405\text{MPa}$, and the resulting limited fatigue life is therefore reasonable.

The final CD//BD sample is sample-h, which exhibited a high fatigue life ($N_f=2.1 \times 10^6$) despite having a relatively large defect size ($\sqrt{\text{area}}=520 \mu\text{m}$), the largest among all tested samples. This behaviour is unusual, as such a large pore defect would typically be expected to significantly reduce N_f . However, as shown in Fig. 8h, this sample is the only CD//BD sample where the pore defect is located far from the sample surface edge. The fatigue crack propagated around the pore and outward through the material, rather than initiating directly from the sample surface (edge). The influence of surface or near-surface defects on stress concentration and fatigue performance in additively manufactured (AM) samples has been well demonstrated [46–49]. It has been reported [46] that the stress concentration factor (K_t) decreases significantly from ~ 4.3 to ~ 2.1 as x/r_d increases from zero to $x/r_d \approx 0.5$, where x is the distance between the defect edge and the sample surface, and r_d is

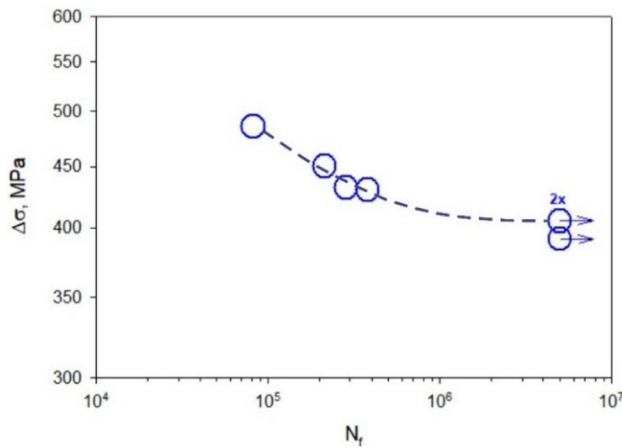


Fig. 11 S–N data for CD//BD showing run-out specimens and failures in which crack growth was not initiated by porosity defects

the defect radius. As shown in Fig. 8h, the CD//BD sample-h has $x/r_d \approx 3$. Therefore, the influence of this defect on N_f is expected to be considerably lower compared with surface or near-surface pore defects observed on the fracture surfaces of other samples.

For the CD⊥BD samples, the data scatter observed in Fig. 6 can be attributed to the pore defects identified on all fracture surfaces (Fig. 9). Crack initiation from the pore defects is clear in Fig. 9a, c, d and e. As shown in these factographs, the cracks propagated directly from the pore defects, forming the fatigue crack growth (FCG) Area 1. All data points are located outside the no-FCG region and its boundaries in Fig. 10, either because $\Delta\sigma > \Delta\sigma_c$ or high $\sqrt{\text{area}}$ values combined with sufficiently high $\Delta\sigma$ levels. Consequently, these data points are positioned away from the no-FCG region and its boundary. In Fig. 9b, crack propagation is observed to radiate from the pore defect. The radiating crack morphology suggests that crack initiation also occurred from the pore defect in the lower region, followed by propagation towards the lower edge of the sample. Although this defect is an internal pore, the x/r_d value is estimated to be close to, but slightly greater than, 1. Due to its internal location, the effect of the defect on fatigue life is reduced. Therefore, despite the relatively high $\Delta\sigma$ (468 MPa) and sufficiently large $\sqrt{\text{area}}$ (218 μm), the sample exhibited a relatively high fatigue life ($N_f = 5.7 \times 10^5$).

Returning to the four CD//BD samples where no pore defects were detected on the fracture surfaces, indicating that pore defects did not initiate the cracks or significantly influence N_f , the corresponding S–N data and run-out results have been re-plotted in Fig. 11. A traced S–N curve has also been proposed in the figure. Although the available dataset is limited, the results show a clear trend towards the endurance limit ($\sigma_e = 405$ MPa, determined from two tests) as the fatigue life increases beyond 1×10^6 cycles. Therefore, the proposed curve can be considered as an upper-bound S–N relationship. This indicates that, in the absence of pore-defect effects, or when such effects are not dominant, the S–N curve shown in Fig. 11 represents the maximum fatigue achievable fatigue strength for the CD//BD samples. The presence of WAAM-induced defects reduces the fatigue strength and fatigue life, resulting in a downward shift of the S–N curve.

As all five CD⊥BD samples failed due to pore defects, a comparable upper-bound S–N curve cannot be established for this orientation. However, a comparison between CD//BD and CD⊥BD samples can be made by examining the fatigue lives (N_f) of three pairs of samples tested at the same stress levels, as summarised in Table 2. For sample-a in both orientations, no pore defects were detected, and failure was not defect-driven; however, the CD⊥BD sample exhibited a higher N_f than the corresponding CD//BD sample. For sample-b, the CD//BD sample shows a $\sqrt{\text{area}}$ value close to $(\sqrt{\text{area}})_c$, while the CD⊥BD sample contains an internal pore. Although both defects are expected to have relatively limited influence, the CD⊥BD sample still shows a significantly higher fatigue life. For sample-c, a large defect is present in the CD⊥BD sample, whereas no defect is observed in the corresponding CD//BD sample. Despite this, the difference in N_f between the two orientations is not substantial, suggesting improved damage tolerance in the CD⊥BD condition. In addition, the endurance limit for CD⊥BD samples (405 MPa), further supports this observation. For CD⊥BD sample-d, although $\sqrt{\text{area}} = 288$ μm , the fatigue life remains high ($N_f = 4.9 \times 10^5$), again indicating strong damage tolerance in the CD⊥BD samples. Overall, these comparisons suggest that the defect-free S–N curve for CD⊥BD samples lies slightly above that of CD//BD. This trend is consistent with the slightly higher ΔK_{th} values observed for CD⊥BD compared with CD//BD samples.

Table 2 Comparisons of $\sqrt{\text{area}}$ and N_f values between of CD//BD and CD⊥BD samples

$\Delta\sigma$, MPa	$\sqrt{\text{area}}$, N_f in CD//BD	Sample	$\sqrt{\text{area}}$, N_f in CD⊥BD	Sample
486	0 μm , 8.2×10^4	a	24 μm , 1.3×10^5	a
468	93 μm , 1.4×10^5	b	218 μm (Int.), 5.4×10^5	b
450	0 μm , 2.1×10^5	c	272 μm , 1.5×10^5	c
435	–	–	$> 5 \times 10^6$	Run-out
405	$> 5 \times 10^6$	Run-out	288 μm , 4.9×10^5	d

4 Conclusions

In this work, the threshold stress intensity factor (ΔK_{th}) of ER70S-6 steel produced by wire arc additive manufacturing (WAAM) was determined through fatigue crack growth (FCG) testing. The ΔK_{th} value for samples with the crack direction (CD) parallel (//) to build direction (BD) is approximately 4.5 MPa $\sqrt{m}$. A slightly higher value of ~ 5.0 MPa $\sqrt{m}$ was obtained for samples with CD normal ($\perp$) to build direction (BD). These are preliminary results as the number of tests is two for each condition. Microstructural and crack path observations suggest that the slightly improved and more localised alignment of the cementite phase in the CD//BD condition may contribute to the reduced ΔK_{th} in these samples. Fatigue life (S–N) testing further showed that the defect-insensitive endurance limit ($\Delta\sigma_{e-h}$) for CD $\perp$ BD samples (435 MPa) is slightly higher than that for CD//BD samples (405 MPa). The ΔK_{th} values obtained in this study were used to evaluate the influence of defect size ($\sqrt{area}$) on fatigue strength ($\Delta\sigma_e$) within the framework of the Kitagawa–Takahashi (K–T) diagram. All $\Delta\sigma\text{--}\sqrt{area}$ conditions associated with crack growth and limited fatigue life (N_f) were found to lie outside the region bounded by $\Delta\sigma_{e-h}$ and $\Delta\sigma_e(\sqrt{area})$, where the latter was determined from ΔK_{th} and $\sqrt{area}$ using Murakami's equation. The presence and distribution of defects, as represented in the K–T diagram, is therefore identified as the primary cause of the observed scatter in the S–N data.

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Data availability No datasets were generated or analysed during the current study.

Declarations

Competing interests The authors declare no competing interests.

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