Evaluation of additive friction stir deposition of AISI 316L for repairing surface material loss in AISI 4340

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Abstract
Additive technologies provide a means for repair of various failure modes associated with material degradation occurring during use in aggressive environments. Possible repair strategies for AISI 4340 steel using AISI 316L deposited by additive friction stir deposition (AFSD) were evaluated under this research by metallography, microhardness, and wear and mechanical testing. Two repair geometries were investigated: groove-filling and surface cladding. The former represents repair of localized grinding to eliminate cracks, while the latter represents material replacement over a larger area, for example, to repair general corrosion or wear. The 316L deposited by AFSD exhibited a refined microstructure with decreased grain size and plastic strain, ~12% lower strength, and 5–12% lower hardness than the as-received feedstock. Wear testing by both two-body abrasion and erosion by particle impingement indicated that the wear resistance of the 316L cladding was as good as, or better than, the substrate 4340 material; however, there was some evidence that the resistance to intergranular corrosion was compromised due to the formation of carbides or sigma phase. In both repair geometries, the microstructure of the substrate beneath the deposited material exhibited heat affected zones where the substrate appeared to have austenitized to a depth of ~4 mm during the deposition process, and transformed to martensite or bainite during cooling. This report constitutes an initial evaluation of a novel approach to the repair of structural steel components damaged by microcracking, wear, or corrosion, with potentially broad applicability to the marine, aerospace, and heavy duty off-road industries.

Keywords Additive friction stir deposition · AFSD · AISI 316 · Repair

1 Introduction
Additive friction stir deposition (AFSD) is a newly developed additive materials technology that is receiving increasing interest for manufacturing applications in a variety of material systems [1–6]. In AFSD, a feedstock material is fed through a rotating tool and axially loaded against a substrate. Because the feedstock rotates with the tool, the axial load applied to the feedstock leads to frictional heating that softens the feedstock and substrate. The feed material extrudes into the space between the rotating tool and the substrate, and concurrent deformation and solid-state mixing occur in a surface layer within the substrate. As a result, a metallurgical bond is formed, without melting, between the filler and substrate alloys [2]. Translation of the tool along the surface generates a layer of deposited alloy, schematically illustrated in Fig. 1a, and subsequent additional passes can be performed to build up material through the deposition of additional layers. AFSD has the potential to provide numerous advantages over other additive processes, including refined microstructure, low final defect fraction, high deposit density, high build rate, adaptability to large-scale application, and low power per pound of build [1, 7].

AFSD is mechanistically similar to the well-known joining technology, friction stir welding (FSW) [8, 9]; however, because FSW does not add material, but simply translates the plasticized material around a non-consumable pin, AFSD provides enhanced capability for volumetric repairs. Known commercially as MELD, AFSD has the potential to enable high-deposition rates during the manufacture of moderately complex geometries [1, 3]. The resultant components often exhibit enhanced wear resistance, corrosion

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protection, and mechanical properties [6, 10, 11]. The first reported applications of AFSD were for manufacturing with aluminum and magnesium alloys [12, 13], and additional results were later reported for Inconel 625 [14, 15], copper [16], Ti-6Al-4 V [17], and aluminum-matrix composites [3]; however, published results for traditional steels are limited [9, 18–20]. Because the solid-state material transfer promotes strong metallurgical bonding while mitigating hot cracking, porosity, heat affected zone (HAZ) formation, and high residual stresses that can be associated with fusion processes, AFSD has obvious potential for repair applications. However, discussion of repair applications for the technology is also largely absent from the published literature [19–24].

This exploratory investigation, which constitutes a first published report of AISI 316L deposited on AISI 4340 steel by AFSD, was performed to assess the potential benefits and challenges in using AFSD for repair of structural steel components damaged by microcracking, wear, or corrosion. Two repair geometries were evaluated: surface-cladding and groove-filling. These geometries simulate replacement of a surface layer to repair general corrosion or wear (Fig. 1a), and repair of deeper grinding to eliminate localized cracks (Fig. 1b), respectively. The substrate material, AISI 4340, is a medium-carbon, low alloy steel of interest for its broad use in applications requiring high strength and toughness, while the AISI 316L filler is an austenitic stainless steel with high corrosion resistance. These materials were selected for marine applications, but the results are relevant to lifetime extension for structural steel components used in a variety of industries, including marine, aerospace, and heavy duty off road.

### 2 Materials and Experimental

#### 2.1 Materials

The AFSD feed material was a solid square rod of AISI 316L, with dimensions of 9.4 mm × 9.4 mm and an average measured hardness of 24 HRC (257 HV0.2). The square profile of the feedstock facilitates transfer of torque from the rotating tool, allowing the tool to drive rotation of the feedstock. Prior to deposition, the feedstock was lubricated with graphite to reduce friction with the tool. The substrate material was AISI 4340 plate with a nominal thickness of 11.5 mm and a specified minimum hardness of 42 HRC (425 HV0.2). Four grooves were machined into the surface of the plate with the dimensions indicated in Table 1. The nominal compositions of the feedstock and substrate alloys are provided in Table 2.

| Groove ID | Met. sample | Depth, mm | Width, mm | Radius, mm |
|-----------|-------------|-----------|-----------|------------|
| 1         | G1          | 0.76      | 4.11      | 3.18       |
| 2         | G2          | 1.52      | 5.44      | 3.18       |
| 3         | G3          | 0.76      | 6.02      | 6.35       |
| 4         | G4          | 1.52      | 8.26      | 6.35       |
Additive friction stir deposition (AFSD)

AFSD was performed by MELD Manufacturing Corporation (Christiansburg, VA, USA) on a MELD B8 system. The deposition used a 39-mm-diameter rotating non-consumable (WC) tool containing a square central channel through which the AISI 316L feedstock was fed. The primary control parameters include the tool rotation rate $\omega$, the feedstock feed rate $V_{feed}$, the tool translation speed $V_{tool}$, and the tool height $z$ relative to the original substrate surface. These parameters, which determine the power and torque, the axial load on the feedstock, the deposition rate, and the layer thickness, are shown in Table 3 along with the calculated torque, axial force, and deposition rate $R_{dep}$.

The process parameters were designed to fill the grooves on the first pass by scaling the tool translation speed $V_{tool}$ and feedstock feed rate $V_{feed}$ to provide a deposition rate $R_{dep}$ (mm$^3$/s) equal to approximately $1.6 \times 2.8 \times$ the volume of the groove and substrate-tool gap transited by the tool each second, $R_{groove} + R_{layer}$ (also mm$^3$/s), where the excess material was applied to ensure complete filling of the grooves and to increase the pressure below the tool in order to maximize shearing of the substrate. These parameters are interrelated via the relationships:

$$R_{dep} \left[ \frac{mm^3}{s} \right] = A_{feed} \left[ mm^2 \right] \cdot V_{feed} \left[ \frac{mm}{s} \right] \quad (1)$$

$$R_{groove} \left[ \frac{mm^3}{s} \right] = A_{groove} \left[ mm^2 \right] \cdot V_{tool} \left[ \frac{mm}{s} \right] \quad (2)$$

$$R_{layer} \left[ \frac{mm^3}{s} \right] = W_{dep} \left[ mm \right] \cdot z_{step} \left[ mm \right] \cdot V_{tool} \left[ \frac{mm}{s} \right] \quad (3)$$

where $A_{feed}$ and $A_{groove}$ are the cross-sectional areas of the feedstock and the groove, respectively, $W_{dep}$ is the width of the deposited material, and $z_{step}$ is the nominal layer thickness as determined by the increment in the $z$-coordinate of the tool between passes. Note that for the 1st, filling pass, the base of the tool was set to a height approximately 0.5 mm above the surface of the substrate. In subsequent passes, $V_{feed}$ and $V_{tool}$ were manually adjusted to maintain the ratio of $R_{dep}$ to $R_{groove} + R_{layer}$ between 1.8 and 2.9.

Three different deposition geometries were executed: single-width cladding, overlap cladding, and groove-filling. The cladding and groove-filling configurations are illustrated schematically in Fig. 1, where the tool translation direction in Fig. 1b is normal to the plane of the image. Note from Fig. 1a that, as the rotating tool is translated, on one side of the tool, the rotation is in the translation direction, and on the other side it is away from it. These two sides are described as “advancing” and “retreating,” respectively. The photograph in Fig. 2a shows the four filled grooves (#1–#4) at the top, and a single-width multipass cladding (SC) at the bottom. Figure 2b shows a region of overlap cladding at the...
2.3 Metallurgical inspection

Samples were removed from the plate by electrical discharge machining, polished to 1 μm diamond suspension, and etched to reveal the microstructure [25]. The AISI 4340 alloy was etched with nital or Vilella’s reagent, while the AISI 316L was chemically etched with Kalling’s reagent or electrolytically etched in 10% oxalic acid. Vickers microhardness testing was performed on a LECO AMH55 with a 200 g load and 13 s hold time.

2.4 Mechanical testing

Tensile and 3-point bend testing were performed using a 100 kN load cell on an Instron 8801 servo-hydraulic testing system. Tensile testing used a gauge length of approximately 12.5 mm and extension rate of 0.085 mm/s, with the rest of the dimensions derived from the guidelines provided by ASTM B557 for rectangular sub-sized specimens [26]. The samples were oriented transverse to the filled grooves, and were cut from the plate by electrical discharge machining. The sample thickness was approximately 0.76 mm for samples T1–T3, and 1.52 mm for samples T4–T6. In addition, a set of tensile samples was machined from an additional single-width multipass cladding (not shown in Fig. 2). Those samples were axially oriented parallel to the AFSD direction, and were comprised solely of the AFSD 316L material.

Bend testing was performed on the section labelled S1 in Fig. 2b with an 85 mm span and a load rate of 74.1 N/s. Load and deflection were monitored during the test, and converted to flexural stress $\sigma_f$ and flexural strain $\varepsilon_f$ by:

$$\sigma_f = \frac{3FL}{2bd^2}, \quad (4)$$

$$\varepsilon_f = \frac{6Dd}{L^2}, \quad (5)$$

In Eqs. (4) and (5), $F$, $b$, and $d$ represent the applied load, sample width, and sample thickness, respectively.

2.5 Wear testing

Two-body wear testing was performed in compliance with the ASTM G174 loop abrasion test (Bud Labs, Rochester, New York 14,616) [27]. The testing used a 100 g load and a 30-μm alumina abrasive tape for a duration of 75 belt passes. Four tests were performed at each location indicated W1–W3 in Fig. 2. For comparison, erosion testing was performed (Bud Labs, Rochester, New York 14,616) in compliance with the ASTM G76 test for erosion by solid particle ($\text{Al}_2\text{O}_3$) impingement using gas jets per the following test parameters: standoff 10 mm, impingement angle 10°, gas
pressure 55.2 kPa, and gas flow rate 118 cm³/s [28]. Three
tests were performed at each of two locations, W1 and W4,
in Fig. 2.

3 Results and discussion

3.1 Characterization of substrate and feedstock

Light optical microscopy (LOM) images of the as-received microstructure in the AISI 316L feedstock are shown in Fig. 3a and b. The microstructure consists of coarse austenitic grains, typically 50–100 μm, that contain twins and shear bands associated with prior cold work equivalent to approximately 10% reduction [29]. LOM and scanning electron microscope (SEM) images of the as-received microstructure in the AISI 4340 substrate are provided in Fig. 3c and d, respectively. The observed microstructure consists of tempered martensite, and is consistent with the nominal hardness of 42 HRC (Hardness Rockwell Scale C).

Table 4 provides the measured yield strength, tensile strength, and hardness of the substrate and filler feedstock, along with specified values related to tempering of the 4340 or cold work in the 316L. The measured Vicker’s hardness of 426 HV0.2 in the 4340 substrate, which corresponds to 43 HRC per ASTM E140-02 [30], is again similar to the specified minimum of 42 HRC. The measured hardness is also consistent with the measured yield and tensile strengths of 1286 and 1368 MPa, respectively, and corresponds to a tempering temperature of approximately 500 °C for 2 h [31]. In the 316L feedstock, the measured tensile strength of 724 MPa approximately corresponds to a 1/8-hard condition per ASTM A666-15 [32].

Fig. 3 Microstructures of the as-received materials: a, b 316L filler alloy after etching with Kalling’s reagent, and c, d 4340 substrate after etching with Vilella’s reagent.
3.2 AFSD groove-filling

Figure 4 shows LOM images of metallurgical cross sections from the four AFSD groove-filled locations G1–G4, indicated in Fig. 2a. The sections have been etched with nital, which reveals the microstructure in the 4340 substrate but has no effect on the 316L filler. The images show the deformation of the groove geometry to be asymmetrical, with more deformation on the advancing side of the tool during the first pass (corresponding to the left side of the images). This is consistent with prior reports of asymmetry in the interfacial mixing during deposition of AA2024 on AA6061, where the asymmetry was attributed to the interaction between the in-plane material expansion (driven by compression of the feedstock) and the material flow (driven by rotation of the tool head) [4]. In addition, the deformation is more pronounced in the smaller-radius grooves 1 and 2 (Fig. 4a and b) than in larger-radius grooves 3 and 4 (Fig. 4c and d). This suggests that the lateral constraint formed by the smaller radius transfers the shear forces to the substrate more effectively. Both of the deeper grooves, Fig. 4a and d, experienced cracking or lack of fusion at the interfaces, indicated by the white arrows in Fig. 4. This is shown in more detail in Fig. 5, where the images reveal different behavior in the two groove geometries. The smaller diameter groove 2 in Fig. 5a exhibits lack of fusion in the deeper parts of the groove (lower right) and well-bonded interface with localized cracking of the substrate in the shallower parts (upper left). In contrast, the larger diameter groove 4 in Fig. 5b exhibits lack of fusion along most of the substrate-filler interface. The feature indicated by the dark arrow in Fig. 5a is a trail of carbide inclusions the AFSD 316L introduced during the stir process.

Beneath the filled grooves, the microstructure shows two distinct layers of HAZ that vary in depth and relative thickness for the different geometries. The microstructures

| Material | Condition | Yield strength, MPa | Tensile strength, MPa | HRB/C | HV | Source |
|----------|-----------|---------------------|-----------------------|-------|----|--------|
| 4340     | Substrate | 1286                | 1368                  | 43C, avg | 426 | Measured |
|          | Temper 205 °C | 1860                | 1980                  | 53C, max | 560 | Ref. [32] |
|          | Temper 315 °C | 1620                | 1760                  | 50C, max | 513 |        |
|          | Temper 425 °C | 1365                | 1500                  | 46C, max | 458 |        |
|          | Temper 540 °C | 1160                | 1240                  | 39C, max | 382 |        |
|          | Temper 650 °C |  860                | 1020                  | 31C, max | 310 |        |
|          | Temper 705 °C |   740                |  860                  | 24C, max | 260 |        |
| 316L     | Feedstock |  586                |   724                 | 24C, avg | 257 | Measured |
|          | Annealed |  205                |    515                | 95B    | -   | Ref. [30] |
|          | 1/8 hard |  380                |    690                | -       | -   |        |
|          | 1/4 hard |  515                |    860                | -       | -   |        |
|          | 1/2 hard |  760                |   1035                | -       | -   |        |

Fig. 4 Cross sections from the metallurgical samples taken from the groove-filling locations defined in Table 1: a G1, b G2, c G3, and d G4. The images show the AFSD 316L fill and overlay on the etched 4340 substrate. The numbers in a indicate the locations of the microstructures shown in Fig. 6: 1, AFSD 316L; 2, HAZ1; 3, HAZ2; and 4, BM. The scale bar in a is valid for all four images
shown in Fig. 6 correspond to the four numbered locations in Fig. 4a: (1) 316L AFSD filler, (2) HAZ1, (3) HAZ2, and (4) substrate. Compared to the as-received 316L microstructure (Fig. 3a and b), the microstructure in the AFSD 316L (Fig. 6a) is significantly refined, with grain size < 25 μm and minimal twinning. The HAZ1 microstructure in Fig. 6b...
is predominantly martensitic, indicating that the region austenized above ~727 °C, and then transformed to martensite during rapid cooling [33]. At the bottom of HAZ1, there is a gradual transition to the HAZ2 microstructure (Fig. 6c), which is comprised of coarse carbides consistent with upper bainite. This indicates that the HAZ2 region austenized, but did not cool rapidly enough to form martensite. In addition, HAZ2 has a significant gradient in the grain size or coarseness of the structure, with the bottom of HAZ2 being finer than the initial tempered martensitic structure found in the substrate. This further supports the idea that HAZ2 austenized during the process, with higher maximum temperatures resulting in larger grains closer to the surface. Finally, the microstructure in the substrate below HAZ2 (Fig. 6d) resembles the as-received structure in Fig. 3c and d. SEM images of the microstructures at the bottom of HAZ2 and the top of the underlying substrate are shown at higher magnification in Fig. 7a and b, respectively. The images confirm the bainitic structure in the HAZ2 to have a finer overall morphology than the substrate, implying a low austenizing temperature, with islands of hypoeutectoidal ferrite.

Figure 8 shows the results of microhardness depth profiling at the locations indicated by the dashed lines in Fig. 4a and b. The data show the hardness of the AFSD 316L to be approximately 88% of the original feedstock, 225 HV0.2 versus 257 HV0.2 from Table 4. This decrease in hardness is associated with dynamic recrystallization that occurs during the AFSD process, leading to the competing effects of relaxation of the strain hardening and decrease in grain size. In HAZ1, the hardness is substantially elevated, exceeding 600 HV0.2 and approaching the maximum quenched hardness of ~670 HV for 0.38% carbon steel [34]. This hardness is consistent with the largely martensitic microstructure observed in Fig. 6b. The HAZ2 hardness ranges from ~347 HV0.2 near the boundary with HAZ1 to ~405 HV0.2 at the boundary with the substrate microstructure, with the gradient being attributed to the observed trend in grain size discussed above. Finally, the hardness in the substrate below HAZ2 ranges from ~340 HV0.2 near HAZ2 to the baseline value of ~426 HV0.2 reported in Table 4. This gradient near the boundary with HAZ1 does not correlate with any visible microstructural changes, and
is attributed to partial stress relieving and tempering that occurs at temperatures between 595 and 675 °C [33]. The hardness data are consistent across the different locations in the two samples, and are summarized in Table 5.

The observed HAZ’s are the result of a complex thermal profile resulting from the rapid heating, short-term exposure at the maximum temperature, and air cooling during each of the multiple AFSD passes. The two zones are differentiated by the maximum temperature attained, which was higher near the surface, and by the cooling rate, which was faster near the surface. These factors interact to affect the extent of austenization, the prior austenitic grain size, and the formation of bainite versus martensite. Similarly, the gradient in the grain size in HAZ2, as well as the refined grain size relative to the original microstructure, suggests that the alloy reached a lower transformation temperature, and cooled less rapidly than HAZ1, with the net effect of converting the original tempered martensite primarily to upper bainite.

The difference in hardness between the AISI 4340 substrate and the AFSD filler metal creates certain implications with respect to mixing at the interface. For example, comparison of the deposition patterns in Fig. 4 with prior published work relating to more evenly matched, and softer, alloys suggests unusually distinct interfaces in the current samples [4, 20]. In fact, prior work on AFSD and FSW has defined several characteristic microstructural zones associated with solid state friction processes: stir zone (SZ), thermomechanical affected zone (TMAZ), heat affected zone (HAZ), and base metal (BM) [7, 8]. The TMAZ refers to regions where the substrate material has been plastically deformed, with evidence of material flow, breakup of inclusions, and distortion of interfaces or boundaries. This region is generally absent from the observed microstructures for 316L AFSD on AISI 4340. This is illustrated in Fig. 4, where the visible deformation of the substrate is limited to localized distortion of the grooves, mostly on the advancing side of the first pass (the left side of the images). As a result of the apparent lack of shear deformation, the extent of mixing of the substrate and filler alloys at the interface was evaluated further.

Figure 9 shows LOM images from the left edge of groove 1 in section G1, and from the right edge of groove 3 in section G3. As with the prior images, the reagent that was used to etch the substrate has no effect on the 316L filler alloy, thus providing a strong contrast between the two materials. The resultant images show evidence of localized mixing, with visible trails of substrate alloy entrained into the filler metal. The observed mixing is typically short-range at the interface, < 50 μm, with some evidence of longer-range entrainment of substrate alloy into the AFSD 316L visible in the upper-half of Fig. 9b. To further reveal this aspect of the mixing, the AFSD 316L filler was electrolytically etched in oxalic acid for 90 s at 1.0 A/cm². This etching condition is designed to reveal possible sensitization due to depletion of Cr from the grain boundaries [35]. The resultant microstructures are shown in Fig. 10 for the area from Fig. 9a. The thickening of the grain boundaries (called “ditching”) in the filler alloy indicated by “1” in Fig. 10b suggests the early stages of potential sensitization, which

### Table 5

|                  | Grooves (Fig. 8) | Overlap clad (Fig. 13) |
|------------------|------------------|------------------------|
|                  | n    | Average | St. dev | n    | Average | St. dev |
| 316 AFSD         | 41   | 225     | 30      | 16   | 249     | 47      |
| 4340 HAZ1        | 9    | 637     | 30      | 20   | 605     | 49      |
| 4340 HAZ2        | 38   | 386     | 29      | 16   | 392     | 17      |
| 4340 Substrate   | 92   | 424     | 16      | 98   | 417     | 11      |

Fig. 9 LOM images showing localized mixing at the interface: a the left edge of groove 1 in section G1, and b the right edge of groove 3 in section G3
can be associated with the formation of carbides due to the ingress of C from the graphite lubricant, or to the formation of sigma phase that has been reported to occur during friction stir processing of AISI 316L [36]. The effect is generally localized to the interfaces between the substrate and filler, indicated by “1” in Fig. 10b and highlighted in Fig. 10c, and between subsequent AFSD passes, indicated by “2” in Fig. 10a. Away from these regions, for example at “3” in Fig. 10b, the etching does not significantly attack the filler alloy, except for a few areas with the appearance of shear-driven flow, location “4” in Fig. 10a. Note that while the localized ditching of the grain boundaries does indicate some degree of decreased resistance to corrosion relative to the unaffected parts of the filler, the oxalic acid etch is a sensitive screening test that may detect conditions which are not necessarily detrimental from a performance perspective [35]. As such, further investigation is required to develop a fuller understanding of the corrosion performance of AFSD 316L, and the potential role of lubricants used during the process.

Tensile testing was performed on sub-sized specimens extracted from grooves 1 and 3, and oriented perpendicular to the AFSD/groove axis. The thickness of the specimens from groove 1, T1–T3, was targeted to match the depth of the groove in order to isolate the contribution of the AFSD filler and the filler-substrate interfaces. In contrast, the thickness of the specimens from groove 3, T4–T6, targeted approximately equal amounts of filler and substrate. The results tabulated in Table 6 show the average yield and tensile strengths for the groove 1 specimens to be 516 and 725 MPa, respectively, while those from groove 3 are 645 and 1391 MPa, respectively.

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Fig. 10 The microstructure from groove 1, section G1, after oxalic acid etching at 1 A/cm² for 90 s to detect sensitization. The areas of b through d are indicated in images a through c. Note that the field of view in b is approximately the same as Fig. 9a.
Table 6 Results of tensile testing AFSD locations perpendicular grooves 1 and 3, with additional data parallel to a thick multi-pass cladding

| Location  | Orientation | Sample ID** | Yield strength, MPa | Tensile strength, MPa |
|-----------|-------------|-------------|---------------------|-----------------------|
| Groove 1  | Transverse  | T1          | 513.7*              | 717.3                 |
|           |             | T2          | 489.5               | 709.6                 |
|           |             | T3          | 542.6               | 747.4                 |
| Groove 3  | Transverse  | T4          | 674.3               | 1345.1                |
|           |             | T5          | 621.2               | 1404.9                |
|           |             | T6          | 640.5               | 1424.3                |
| Cladding  | Axial       | N/A         | 591.6               | 634.2                 |
| 316L only |             | N/A         | 405.4               | 695.2                 |
|           |             | N/A         | 561.3               | 618.4                 |

*Estimated from cross-head displacement due to malfunction of extensometer
**From Fig. 2
***Not shown

Figure 11 shows the fracture cross sections for samples T1 from groove 1 and T5 from groove 3. The images show ductile necking in the AFSD filler (top), with minimal deformation in the harder substrate (bottom). The T1 sample is shown in Fig. 11a, to be predominantly filler alloy (~80%), while the T5 sample is approximately one-half substrate (~50%), which accounts for the elevated strength values. Assuming a linear rule of mixtures, a simple linear regression can be used to estimate the tensile strength of the 4340 substrate and 316L filler as 1599 and 524 MPa, respectively. For the 316L filler, this corresponds to a well-annealed condition per Table 4, which is consistent with the microstructure observed in Fig. 6. For the 4340 substrate, the interpretation is more challenging due to the non-uniform microstructure discussed above. However, the data indicate that immediately below the grooves, the substrate material has an elevated tensile strength, which is consistent with the elevated hardness and the observed martensitic microstructure. In both samples, the fracture propagated through the filler alloy, rather than along the interface with the substrate, and occurred at approximately the minimum thickness of the underlying substrate material. This indicates that, for this test geometry, the interfaces are at least as strong as the filler alloy. For comparison, an additional set of tensile tests was performed on samples extracted from a thick single-width cladding, and oriented axially along the AFSD deposition direction. Those samples, which were comprised solely of AFSD alloy with no attached substrate material, had average yield and tensile strengths of 519 and 649 MPa, respectively.

3.3 AFSD cladding

Surface cladding by AFSD was also explored as a potential method suitable for material replacement over larger areas. Figure 2b showed a region of cladding that was deposited with approximately 20% overlap between passes. Four metallurgical samples prepared from the locations identified as O1–O4 in Fig. 2b are shown in Fig. 12. The cross sections show a similar HAZ pattern comprising two layers with microstructures comparable to those previously shown in Fig. 6. Hardness depth profiles performed at the locations indicated by the dashed lines in Fig. 12c and d are presented in Fig. 13. The depth profiles are comparable to those from the groove-filling geometry presented in Fig. 8, with a slight decrease in the hardness of the filler material relative to the as-received condition, HAZ1 hardness again in excess of 600 HV0.2, HAZ2 hardness averaging 392 HV0.2, and BM hardness increasing from a minimum near HAZ2 to the baseline measured in the as-received plate. In general, HAZ1 is thicker than in the groove-filling examples from Fig. 8, while HAZ2 is less pronounced and lacks the measurable gradient between the boundaries with HAZ1 and BM. These differences are attributed to the different thermal histories associated with the two substrate geometries and AFSD process parameters. The hardness data are again summarized in Table 5.
The sharp appearance of the interface between the substrate and cladding shown in Fig. 12 and the absence of a significant TMAZ below the deposited material are similar to the characteristics observed in the groove-filling geometry (Figs. 4 and 6). Closer inspection of the interface reveals comparable localized mixing (Fig. 14a and b) and evidence of sensitization of the associated grain boundaries (Fig. 14c and d). Additional etching is visible in the interface between consecutive passes and along apparent flow lines (Fig. 14b).

As above, the etch-pattern suggests that the shear mixing that bonds the interface has a potentially deleterious effect on the localized corrosion resistance of the filler alloy, possibly due to the formation of carbides enabled by incorporation of C from the graphite lubricant.

The subsurface hardness of the AFSD 316L cladding was shown in Table 5 and Fig. 13 to be ~ 249 HV0.2, which is similar to the normal surface hardness of 96 HRB determined by Rockwell testing. In contrast, the measured subsurface hardness of the as-received substrate provided in Table 4 is 426 HV0.2, while the normal surface hardness determined by Rockwell testing is 41 HRC. The difference in hardness between the as-received AISI 4340 substrate and the surface of the AFSD 316L filler material may raise some concern about loss of wear resistance resulting from resurfacing with 316L deposited by AFSD. This was evaluated by performing two types of wear testing at the locations indicated W1–W4 in Fig. 2b: two-body wear by abrasive loop contact and erosion by solid particle impingement [27, 28]. The results are provided in Table 7, from which it can be seen that the volume loss in the AFSD 316L is less than 70% of that in the as-received 4340 substrate during the two-body abrasion, regardless of the difference in hardness. The results also show that the mass loss during the erosion test was comparable for the two materials, differing by less than 3%.

Studies have shown a linear relation between hardness and abrasion resistance for steels with similar metallurgical phase compositions; however, recent work has also shown a strong dependence between wear resistance and phase composition during two-body abrasion of steels with the same initial hardness [37, 38]. Narayanaswamy et al. investigated four steels with different microstructures, but all with hardness in the range of 325–360 HV, and reported wear rates in the following order: pearlite < bainite < martensite < tempered.
martensite [37]. In that work, for an abrasive (SiC) particle size of 58 μm, the relative two-body wear rates of the four phases were found to be 1.00, 1.36, 2.59, and 2.62, respectively. The differences in wear rate were attributed, in part, to the capacity of the two-phase microstructures (pearlite and bainite) to accommodate realignment of the harder carbide phase to the abrasion sliding direction by ductile deformation of the softer component. In addition, the lack of plastic

**Fig. 14** LOM images of the interface from overlap clad section O3: **a** at the location indicated by the dashed box in Fig. 12c, prior to electrolytic oxalic acid etch, **b** the same location at higher magnification after oxalic acid etch, **c** along the interface with HAZ1 after oxalic acid etch, and **d** SEM image taken after oxalic acid etch at the location indicated by the black arrow in **b**

| Material | Test location* | ASTM method | Number of tests | Average volume loss, mm³ | Average mass loss, mg |
|----------|----------------|--------------|-----------------|--------------------------|------------------------|
| 4340     | W1             | G174**       | 4               | 1.582                    | N/A                    |
| 316L AFSD| W2             | G174**       | 4               | 1.086                    | N/A                    |
| 316L AFSD| W3             | G174**       | 4               | 1.069                    | N/A                    |
| 4340     | W4             | G76***       | 3               | N/A                      | 9.4                    |
| 316L AFSD| W4             | G76***       | 3               | N/A                      | 9.2                    |

*From Fig. 2

** Abrasion resistance by abrasive loop contact

***Erosion by solid particle impingement using gas jets
deformation in the martensite and tempered martensite makes those phases more vulnerable to Hertzian fracture on the abrading surface, potentially leading to higher wear rates [39]. These findings can be applied to the observed performance as follows.

It is well established that austenitic stainless steels undergo strain-induced martensitic transformation in a near-surface layer during abrasive wear [40–42]. This creates a multi-phase microstructure that allows realignment of the harder martensite towards the abrasive sliding direction, and favors ploughing and wedge formation with narrow/deep surface tracks. In contrast, the tempered martensite of the AISI 4340 resists plastic deformation, thus favoring cutting (material loss) with wide/shallow surface tracks. Figure 15 shows SEM images of the wear surface from the as-received AISI 4340 and the AFSD 316L, locations W1 and W2 in Fig. 2b. The images show deeper grooves with more plastic deformation and obvious ploughing in the AFSD material, Fig. 15b. In contrast, the surface of the AISI 4340 in Fig. 15a exhibits shallower grooves with less evidence of plastic deformation, which is consistent with cutting as the dominant abrasion mechanism. In this context, it is noteworthy that the mechanisms involving plastic deformation during abrasive wear, ploughing, and wedge formation, only deform/translate material, whereas cutting results in material removal. The current results for the two-body abrasive loop test, Table 7, are thus consistent with prior reports of elevated wear resistance in steel microstructures comprised of two phases, hard and soft (brittle and ductile), relative to microstructures comprised of a single hard phase. In contrast, sub-surface investigations on polished 316L suggest that particle impacts result in the formation of a layer of strain-induced martensite [43], which explains the similarity relative to AISI 4340 in wear rate during solid particle impingement.

Three-point bend testing was performed on the multi-layer cladding designated S1 in Fig. 2a, and for comparison on bare 4340 substrate. Figure 16a shows the test configuration, where the span $L$ is approximately 90 mm and the deflection $D$ is such that the AFSD 316L is on the tensile side of the sample. Figure 16b shows plots of the flexural stress $\sigma_f$ versus flexural strain $\varepsilon_f$ for the bare and clad substrates. The results, which are summarized in Table 8, show that the yield and ultimate strengths of the clad sample are 60% and 73% of the unclad substrate, respectively. This is due to the lower tensile strength of the AFSD 316L relative to the 4340 substrate. The clad bar also shows a 47% improvement in the strain to failure, which is again attributable to the performance of the 316L cladding on the tensile side of the bar. Note from Table 8 that the samples were configured to be similar in thickness, 11.3 vs. 14.1 mm, but with approximately one-half (52%) of the clad sample comprising cladding. Thus, the interface between the cladding and substrate is approximately at the neutral axis, neglecting the slight difference in moduli, with the cladding in tension and the substrate in compression. Figure 16c shows the clad sample after the bend test. Near the bottom of the image, it can be seen that the final AFSD layer has fractured and delaminated. This indicates poor interlayer bonding at this thickness of cladding (7.4 mm).

### 3.4 Selection of the AFSD process parameters

The process control parameters defined above, $\omega$, $V_{feed}$, $V_{tool}$, and $z$, determine the energy input to the interface in the form of frictional heating and shear deformation. Excess input heat (high $\omega$, high $V_{feed}$, and/or low $z$) increases the depth and severity of the HAZ, and increases the risk of warping or cracking of the substrate. Insufficient heat input leads to surface galling and/or poor interfacial or interlayer bonding [44]. In order to expedite this initial process assessment, the process parameters were not actively optimized, but rather were selected based on prior experience at MELD. The results shown above demonstrate that the selected parameters achieved sufficient heat input for successful interlayer bonding without surface galling. Poor interfacial bonding was observed in some groove locations; however, this effect appears to be related to the groove depth and geometry, and not solely to the input heat. As such, future process optimization for repair applications should explore groove
4 Summary and conclusions

Possible repair strategies for AISI 4340 steel using AISI 316L deposited by AFSD were evaluated by metallography, microhardness, and wear and mechanical testing. The materials were selected based upon interest for marine applications, but the results are relevant to lifetime extension for structural steel components used in a variety of industries, including marine, aerospace, and heavy duty off road. Compared to the as-received 316L, the microstructure in the AFSD 316L was significantly refined, with decreased grain size and no residual evidence of plastic deformation. Microhardness testing indicated a slight decrease in the hardness after AFSD, 4–12%, while tensile testing indicated a comparable strength decrease of ~12%. Wear testing by two-body abrasion and erosion by particle impingement both indicated that the wear resistance of the AFSD 316L was as good as, or better than, the substrate 4340. Some evidence was observed to suggest that the resistance to intergranular corrosion was compromised in the AFSD 316L, whether due to diffusion of C from the graphite lubricant or to formation of sigma phase.

Two repair geometries were investigated: groove-filling and surface cladding. The former represents repair of localized grinding to eliminate cracks, while the latter represents material replacement over a larger area, for example, to repair general corrosion or wear. In both repair geometries, the microstructure of the substrate beneath the deposited material exhibited upper and lower heat affected zones. The appearance and microhardness profiles of these zones suggest that in both HAZs the substrate reached a sufficiently high temperature to transform to austenite. In the upper HAZ, the subsequent cooling rate was sufficiently rapid to transform the austenite to martensite, while in the lower HAZ, a slower cooling rate resulted in transformation to upper bainite. Below the lower HAZ, the temperature was never high enough to transform the material to austenite, and the substrate retained the original tempered martensite microstructure. An observed gradient in the grain size in the lower HAZ, as well as the refined grain size relative to the original microstructure, suggests that the maximum (transformation) temperature in the HAZs decreased with depth during the AFSD process.

This investigation of AISI 316L deposited on AISI 4340 steel by AFSD was performed to provide an initial assessment of the potential benefits and challenges AFSD for repair of structural steel components damaged by microcracking, wear, or corrosion. While the results presented above are generally promising, several opportunities for additional investigation were identified. Future work should evaluate the impact of the martensite in the upper HAZ upon the fracture toughness, and optimize the process parameters.
to minimize the effect. Additionally, the effect of groove geometry should be further evaluated in order to maximize repair efficiency. Finally, while the AFSD method does not introduce any keyhole effects at the start-stop locations, as occurs with FSW, the effect of the different thermal history at these locations should be evaluated.

**Author contribution** The authors confirm contribution to the paper as follows: study conception and design: AL and MW; data collection: LPM and AL; analysis and interpretation of results: LPM and AL; writing the first draft manuscript: LPM. All authors reviewed the results and approved the final version of the manuscript.

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