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ABSTRACT

A robust method is introduced to simulate and study the filler wire metallurgy for controlled cooling conditions after melting, enabling efficient mapping with prompt analysis of trends. Proposed is a reduced, though representative, process with more controllable conditions. Short lengths of filler wires are preplaced in a cavity, drilled into a base metal sheet. Irradiation by a pulsed laser beam melts the wire to generate a sample nugget. Pulse shaping influences the cooling rate, granting the ability to tailor weldament microstructures. The method is demonstrated for S1100QL steel and undermatched filler wire, to obtain high toughness for processes like laser-arc hybrid welding, where a representative thermal cycle is needed. For high toughness, a controlled amount of acicular ferrite and, in turn, nonmetallic inclusions is desirable. This “snapshot” method has revealed a characteristic histogram of inclusion sizes, for different pulse shapes. Additional information on the thermal cycle can be acquired by employing thermocouples, a pyrometer, or advanced methods like high speed imaging or modeling. The method offers a wide spectrum of variants and applications.

Key words: filler wire, consumable, welding, thermal cycle, microstructure, cooling rate

I. INTRODUCTION

Microstructures of processed metals are a determining factor of mechanical properties in manufacturing techniques, particularly in fusion processes exemplified by welding. For certain welding techniques, filler material is used to bridge gaps, which can have a unique chemistry independent of the base material, causing compositional changes. Effects in the microstructure are generally observed postprocessing, after addition of alloys or thermal treatments. For welding processes, weld metal microstructures and the heat affected zone (HAZ) can have large compositional changes over marginal lateral distances due to local thermal histories, presenting a challenge to identify the mechanisms and causes for microstructures in specific regions.

The Gleeble-method enables the replication of the HAZ of welded joints, providing full samples with uniform microstructures, under controlled thermal cycles. Newer Gleeble systems are capable of simulating solidification processes, with cooling rates up to 200 °C/s from 800 to 500 °C (the critical cooling interval that determines specific microstructures, also expressed by the duration t8/5) in a 10 mm plain carbon steel bar with 10 mm free span reduced to 6 mm diameter for 6 mm long at midspan. By quenching measures, up to 10 000 °C/s can be achieved at the surface, then inhomogeneously. However, Gleeble tests are done on homogeneous pieces or easily joined samples (square bricks, cylinders) but not for filler wires. Exceptionally, a powder sample was used in a simulated forging process, by compression in a Gleeble 3800 system, after 3D-printing a near-net shape cylindrical test sample. Moeinifar et al. simulated a double pass thermal cycle in the HAZ, comparable to submerged arc welding heat inputs, having an initial peak temperature of 1400 °C followed by a reheating to 800 °C with varied cooling rates on X80 pipeline steel in a Gleeble system. This was done to measure the size and area fraction of martensite/austenite constituents with a field emission scanning electron microscope, SEM. Fu et al. observed microstructural changes under Gleeble-induced thermo-simulation.
compressions of hybrid deposition and microrolling cylindrical samples. The sample was produced through depositing multiple layers of filler material into a wall structure. Impact toughness was reduced through the enhancement of intensity direction of deformation textures. However, the wire material already experiences a thermal treatment before testing in the Gleeble system, hence not microstructures directly from melt.

Similar to Gleeble is the Jominy end-quench test, for which the cooling cycle was measured by thermocouples. Occasionally, sophisticated approaches to study the formation of microstructure originating from the molten state are applied. For example, preparation of a homogeneous melted sample in a furnace followed by heat treatment along with in situ observation of the transient metallurgy during cooling by a high temperature-laser scanning confocal microscope (recording at 15 fps), or by a dipping technique of a ceramics-covered steel substrate into a larger melt.

Wire filler processes like laser-arc hybrid welding or narrow gap multilayer laser welding may not efficiently be demonstrated in this manner. Linking resulting metallurgical compositions and distributions to their origins requires isolation of the mechanisms; a challenging feat with significant variations over modest lateral distances with a local thermally and chemically dependent characteristic metallurgical domain. With complex heat inputs from these processes, tailoring of the microstructure is possible through prolonged cooling, special chemistry wire, and dilution of base material or through multiple heat cycles, which can lead to distinct metallurgical compositions and distributions. The most established method to characterize filler wires is the generation of multiple gas metal arc welding tracks, wide enough that the weld metal can be extracted for mechanical testing, typically tensile testing or Charpy V-notch testing, CVN. This method does not provide information on the local link to the thermal cycles and microstructures. Moreover, the wire test sample is composed of inhomogeneous melting and heating cycles.

The thermal cycles experienced by a cell of metal during welding can to a certain extent be measured by employing thermocouples, by pyrometers, or by thermal imaging. Recently, advanced thermal imaging techniques have been developed and more accurate identification of the emissivity was achieved, for calibration. While most of the measurements are limited to surface temperatures, numerical simulation aims to predict the volumetric temperature field, which is altered by convection from complex melt pool flow. Apart from Marangoni-convection, depending on the welding technique additional driving forces can govern the flow, for laser beams particularly ablation pressure in the case of keyhole boiling. For the here presented method, which is based on laser spot melting in conduction mode, thermocapillarity flow is the main driving force, which was studied by computational fluid dynamics (CFD) (Ref. 9) and by high speed imaging (HSI), apart from evidence resulting from altered weld cross section shapes. The melt flow is here also of importance with respect to inhomogeneity of dilution of the wire with the base metal. To a certain extent, for laser spot welding, the influence of different pulse shapes on the thermal cycle and on the resulting microstructures has been investigated, usually combinations of ramping up or down laser power, for example, for dissimilar metals in dental applications.

The here presented method, to study and tailor the microstructure of filler wires, was initially developed to improve the toughness of suitable weld metal for Q+T steels up to 1100 MPa yield strength, for which the method will here be demonstrated. For HSLA Q+T steels like Domex 960, strengthening precipitates like carbides heavily influence the requirements and specifications of welding procedures. A high cooling rate, \( t_{\text{8/5}}\) of 0.6–1.3 s, from low heat input provided a fine grained microstructure composed of upper and lower bainite, martensitic islands, polygonal ferrite, and allotriomorphic ferrite. For Strenx 1100 MC steel, a decrease of tensile strength was seen compared to the base material, seeing even only 50%–60% of the impact toughness.

One promising contribution to high toughness is acicular ferrite, AF, which is likely to develop instead of bainite during cooling of microstructures with a local thermally and chemically dependent characteristic metallurgical domain.
nonmetallic inclusions, NMIs.\textsuperscript{13} 10\%-36\% of inclusions were identified to initiate nucleation of AF.\textsuperscript{14} A minimum inclusion size of 0.2 \(\mu\)m was sufficient to decrease the free energy barrier for solute depleted zone nucleation of AF.\textsuperscript{15} Cooling rates of 20–30 \(\circ\)C/s were needed to change the microstructure from polygonal ferrite or allotriomorphic ferrite to fully acicular microstructures, or a mixture of AF and Bainite.\textsuperscript{16,17} Mo addition enhanced AF formation, providing a dual lath morphology of dominant parallel laths and interwoven/chaotic ones to a lesser extent.\textsuperscript{18} A two stage cooling cycle allowed for high volume fraction of AF with lower cooling rates.\textsuperscript{19}

Ti-inclusions inhibited the competitive formation of allotriomorphic ferrite, polygonal ferrite, and Widmanstätten ferrite. A decrease of the free energy barrier around the inclusion, increasing lattice disregistry was the cause.\textsuperscript{15,18} An increasing potential to Ti-oxide, Ti-oxysulfide, and then Ti-Ca-oxysulfide inclusions was listed after addition of sulfur or calcium to control austenite grain size and inclusion density.\textsuperscript{16} TiO and TiN inclusions are potent, low lattice mismatch nucleants. However, MnS conversely was considered an inactive nucleant unless B was added. B, which when segregated to the prior austenite grain boundaries, enhanced the activation energy barrier.\textsuperscript{19} An increase of Ti had a beneficial effect on AF nucleation, noting a limitation of long-range diffusion of Mn, leading to the Mn-depleted zone around the NMI that was seen as a driving force by other groups.\textsuperscript{20}

In order to study and improve filler wires in more detail with respect to the microstructure linked to the thermal cycle and chemistry, a high productivity testing method is here presented for the first time. Pulsed laser melting of preplaced wire enables simpler and more controllable conditions, while maintaining a likeness to the original welding process. The method is demonstrated for the growth of nonmetallic inclusions for an unmatched filler wire for S1100L Q + T steel as the base metal.

II. METHOD

A new approach is here presented to study microstructure evolution through tailored thermal cycles, in a high productivity technique that is representative of wire-based welding, through laser pulse shaping.

Cavities were drilled into 12 mm thick, grain refined S1100QL steel sheet, diameter 5.5 mm, and a central depth of 7 mm, represented in Fig. 1. Low alloy solid filler wire designated LNM-MoNiCr (from Lincoln Electric, according to EN ISO 16834-A: G Mn4Ni2CrMo), diameter 1.2 mm, was cut to lengths of 8 mm. Chemical and mechanical properties of the base and filler material are shown in Tables I and II. Sixteen cleaned wire lengths were vertically placed into each cavity, slightly sticking out.

The laser beam, centrally aligned in the cavity, was produced by a 1070 nm wavelength Yb-fiber laser (IPG YLR-15000, 15 kW cw, can be pulsed and pulse-shaped to >200 \(\mu\)s-steps, beam parameter product, BPP, 10.3 mm mrad, feeding fiber diameter 200 \(\mu\)m). The laser beam then passes the beam switch to a 400 \(\mu\)m diameter processing fiber (BPP 14.6 mm mrad) before being focused by the optics (Precitec YW52) and projected to the base material surface. The collimating and focusing lenses had focal lengths of 150 and 250 mm, respectively. The beam was defocused by setting the focal point 12 mm above the base material surface, giving a spot size of approximately 1.5 mm. The optics were inclined 10\(^\circ\)-15\(^\circ\) to prevent back reflections. The shielding gas was 18 l/min Ar with a second cross jet 5 mm above the shielding gas tube.

The pulsed laser beam can be modified with respect to power level, duration, and stepping/ramping. Of the manifold pulse types that have already been studied, results for three pulse types are here presented, combining square pulses and ramping down, as will be shown later.

Measurements of the thermal cycle were conducted in a variety of different setups. A type-K thermocouple probe (max. 1260 °C) was used to measure the surface temperature at a certain distance from the edge of the cavity. A distance of 3 mm corresponded approximately to the center of the infrared emission field of a pyrometer (max. 900 °C, 6 mm spot size at focus). The pyrometer was calibrated to a probe; an emissivity of 0.80 corresponded to oxidized steel, since the base plate was manually abraded to remove the majority of oxides. As another option, a sacrificial thermocouple wire was inserted through a channel from the bottom of the cavity (so that the melt could drop onto the contact). Secondary experiments were employed with rodlike thermocouple probes and sacrificial thermocouple wires, diameter 0.5

### TABLE I. Chemical composition (wt. %)\textsuperscript{a} of the here studied base metal, S1100QL, and wire (balance: Fe).

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Cu</th>
<th>Ni</th>
<th>Mo</th>
<th>B</th>
</tr>
</thead>
<tbody>
<tr>
<td>S1100QL</td>
<td>0.21</td>
<td>0.5</td>
<td>1.4</td>
<td>0.02</td>
<td>0.005</td>
<td>0.8</td>
<td>0.30</td>
<td>3</td>
<td>0.7</td>
<td>0.005</td>
</tr>
<tr>
<td>LNM-MoNiCr</td>
<td>0.09</td>
<td>—</td>
<td>1.8</td>
<td>—</td>
<td>—</td>
<td>0.3</td>
<td>—</td>
<td>2.20</td>
<td>0.55</td>
<td>—</td>
</tr>
</tbody>
</table>

\textsuperscript{a}Intentional alloying elements.

### TABLE II. Mechanical properties of the base metal, S1100QL, and wire.

<table>
<thead>
<tr>
<th>Plate thickness (mm)</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation A (%)</th>
<th>Impact strength (J)</th>
</tr>
</thead>
<tbody>
<tr>
<td>S1100QL</td>
<td>5–40</td>
<td>1100</td>
<td>1250–1550</td>
<td>10</td>
</tr>
<tr>
<td>LNM-MoNiCr</td>
<td>—</td>
<td>&gt;890</td>
<td>950</td>
<td>&gt;15</td>
</tr>
</tbody>
</table>
mm, inserted from the top of the cavity in a gap between the wires. Concurrently, HSI was used at 500 fps with a declined view onto the cavity, recording the thermal process radiation. This was for collaboration information.

The weld nuggets were produced so that they were easily replicated. Laser placement influenced the shape of the top of the nugget and would only influence dilution if an off-center placement was used. Fusion, and hence dilution, to the base material was due to either off-placement of the laser or extended exposures. The melt nugget samples were cut, longitudinally, just off center so that the polishing process would remove the remainder so that the center of the weld was exposed. Standard metallographic procedures were used to polish the melt nuggets to a mirror surface before etching with 2% Nital. The microstructures of the melt nuggets were imaged using two different optical microscopes, an Olympus VANOXRT and a Nikon Eclipse MA200, which provided different visualization.

NMIIs were evaluated in etched samples through an image processing software, FIJI (altering contrast and brightness, and applying median filters), to count the number of inclusions as well as to determine the distribution of sizes (areas) of NMIIs and their number density.

The analysis of the thermal cycle and the study of laser pulse shape options were supported by a simple mathematical model. An analytical solution of the temperature field $T$, Eq. (1), along the vertical axis $z$ as a function of time $t$ is

$$T(x = 0, y = 0, z; t) = T_a + \frac{A_0 D}{\kappa} \left[ \text{erfc} \left( \frac{z}{D} \right) - \text{erfc} \left( \frac{\sqrt{z^2 + w_0^2}}{D} \right) \right]$$

with the thermal penetration depth $D$ given in the following equation:

$$D = 2\sqrt{\kappa t},$$

where $w_0$ is the circular laser pulse of radius, $I_0$ is the power density (square profile) and $t_L$ is the duration at the surface of a semi-infinite solid, $K$ is the thermal conductivity, $\kappa$ is the diffusivity, $T_a$ is the ambient temperature, and $\text{erfc}(\cdot)$ is an integral complementary error function. The average absorptance $A$ needs to be estimated (e.g., $A = 20\%$). One then needs to superimpose a second, identical pulse at $t_L$, as shown in the following equation:

$$T(z; t) = T_a + T_{ON}(z; t) - T_{OFF}(z; t - t_L).$$

Any additional pulses can then be superimposed, to design or approach almost arbitrary temporal pulse profiles. The start and end time of each pulse as well as the pulse power level needs to be considered, for each pulse, or pulse-step. The model bears a number of simplifications (like constant material properties independent of temperature, no melt flow, no latent heat of melting/solidification, uncertain absorptance, lacking surface heat losses, etc.) but is suitable for basic understanding or fast estimations.

III. RESULTS

In the following, the new snapshot method is presented through results and observations demonstrated for the wire and base metal given in Table I. Figure 2 shows cross sections of the melted wire for strong dilution, moderate dilution (asymmetrically, because of beam misalignment), melting of the top region only (hence the wire pieces and cavity can be well seen), and sole melting of the wire without dilution (this is a parameter-robust option, to just melt and quench the whole wire). The arrows point to pores.

Additionally, beam focus and irradiation duration have an effect on the dilution. Limiting radiation and melting to the filler material allows for evaluation of just the wire chemistry and cooling rate. With dilution of the base material, other elements not present in the filler material could influence the final microstructure.

The micrograph in Fig. 3(a), magnified in Fig. 3(b), shows hard microstructures of martensite and bainite accompanied by
softer microstructures like Widmanstätten ferrite and polygonal ferrite. Some of the ferrite structures observed were in contact with an NMI, but not in an obvious manner to exclude other microstructures in favor of AF. The lack of a larger percentage of AF would indicate that the cooling rates were too high. Pores (upper blue arrow) and NMIs (lower orange arrow) can be seen, which are surrounded by a halo of bright phase, a region that represents an elemental depletion zone. The majority of the inclusions are found in the light phase (polygonal ferrite) either being encapsulated or central to a number of grains. Figures 3(c) and 3(d) show the differences in NMI size between the weld metal center and edge, respectively. Here, the Olympus microscope provided a clearer view on the NMIs while the above images, made with the Nikon microscope, had better visibility of the microstructures. The larger NMIs tend toward the grain boundaries while the smaller NMIs are found in the center of the grain. Pores and inclusions were observed in all pulses. The microstructures as well as the NMI number density and their distribution depended on the heat input and pulse duration.

Figure 4 shows the method to count NMIs by sectioning of a micrograph, then selecting a threshold for NMIs, and filtering the NMIs from noise. From Fig. 4(c), the areas of the NMIs can be measured and counted, as a histogram. While this example is shown for NMIs, it can similarly determine the area fraction of the microstructural constituents, for which the grain boundaries are shown in Fig. 4(b).

Figure 5 is a series of HSIs, providing additional information on the process. The first seconds of HSI, Fig. 5(a), show the wires beginning to melt from the center, while the outer wires can be seen to be still solid. Later, Fig. 5(b), also the outer wires are molten. A vapor flow can be seen that disappears at a later stage, Fig. 5(c), and makes the melt surface well visible because the laser power is ramped down. After switching off the laser pulse, the melt nugget cools down, which can be seen from decreasing brightness, see Fig. 5(d). From such HSI, for example, the solidification front speed can be measured. While the here shown HSI originates from thermal process emissions, by employment of an illumination laser along with a narrow band emission blocking filter, different images with further information can be obtained.

Figure 6 is a histogram of the NMI counts, from the shown three different pulse profiles A, B, and C, taken over the same area. Preliminary parameter studies showed that ramping the power down avoids quenching and was thus studied further as longer
cooling times are of more interest in respect to NMIs. In addition, the model confirmed that ramping down enables guided temperature decays, at least close to the surface. For each pulse, there is a decrease in the number of NMIs as the size of the NMIs increases. However, disregarding the underflow and overflow bins, there is a peak in all three profiles at 0.61 μm² (or an average diameter of about 900 nm), bearing in mind that visible sections are normally at a random location, smaller than the inclusion as a whole. For these conditions, the NMIs preferably grow to this size, robust against different pulse duration and power level among the three pulses. Both pulse A and pulse B have a significant quenching step going from a laser power of 1.5 kW to having no laser emission. Pulse A has an extra second of exposure at the lower power. As there is a gradient of the micro-structures and NMIs to the exterior, it is plausible that they are not depicted in this limited two-dimensional micrograph. Slight offsets to the center of the melt nugget could also be the reason for the changes in the NMI count.

Pulse C has a more gradual decline than the previous pulses, having a less dramatic cutoff of laser power at 800 W. Pulse A caused finer grains than pulse C. Clustering of smaller inclusions to form larger NMIs can take place which would be consistent with Ostwald ripening, the positive correlation of heat input to NMI size.

Apart from HSI, the effects of the heat input can be recorded through measuring the temperature at known locations outside the process zone. Figure 7 is a depiction of the thermal curves recorded by three different types of measuring devices, at different locations: a type-K thermocouple probe, a sacrificial type-K thermocouple wire, and a pyrometer, alongside calculated values. The steps in the curves are due to the sampling rate of the equipment. The thermal cycles are steeper closer to the center, which is confirmed by the calculations that also explain a delay of the peak with increasing distance. Although the measurements and calculations provide useful accompanying information, they bear uncertainties to be reduced.
The snapshot method demonstrates a new technique to rapidly test many different variables through a wide variety of wire and base metals, including even testing of powder.

Limitations of the method, to be carefully considered or overcome (e.g., supported by numerical simulation), are limited thermal conductivity through the cavities between the wire pieces, uncertain absorptance particularly initially on the solid wire pieces, inhomogeneous material merging, and yet thermal gradients in space.

From pulse-tailoring, desired thermal cycles can be identified, as a guideline to correspondingly improve the welding process, e.g., by postheating or by suitable time and space management of multiple thermal cycles, e.g., in narrow gap multitaper laser welding or even in additive manufacturing. An example of the creative potential is to generate by a first pulse a well-defined nugget with controlled dilution while a second pulse can independently induce the desired thermal cycle in the solid nugget. Because of its simplifications, the method provides good access for various measurements and supportive accurate CFD simulation is realistic, at least once the wire is molten. The advantages of laser beams are made use of, as a highly controllable tool with respect to energy, space and time.

From systematic mapping, a gradual improvement in the understanding of the links between thermal cycles, wire chemistry, and obtained microstructure can be expected, in an efficient manner because of the high productivity of the method. Based on this understanding, optimization can be aimed at. Accompanying statistical evaluations can identify and improve the robustness, eventually for welding (or other processes). From improved wire chemistry and welding processes, through the microstructure improved mechanical properties are aimed at.

IV. CONCLUSIONS

(i) A new filler wire testing method was developed that enables one to isolate mechanisms, to tailor the thermal cycle and microstructures by pulse shaping, representative for continuous welds, and to rapidly map parameter variations.

(ii) Laser parameters were identified that provide good control over the cooling cycle, dilution, and microstructures; wire nuggets can even be generated without dilution.

(iii) The snapshot method has demonstrated that a characteristic size distribution of nonmetallic inclusions is formed for specific pulse-ramping.

(iv) The method offers a wide spectrum of variants and applications, including additional experimental and theoretical information.

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REFERENCES


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Stephanie Robertson was born in Syracuse, New York, in 1992 and is currently employed as a Ph.D. at Luleå University of Technology (Luleå, Sweden). In 2016, she received a dual master’s degree from LTU and from Polytechnic University of Catalonia (Barcelona, Spain) in Material Science, specifically nanomaterials and biocomposites, while her bachelor’s diploma is from Clarkson University (Potsdam, NY) in Chemical Engineering with a focus in Biomedical Engineering.

Jan Frostevarg is an associate professor at Luleå University of Technology where he gained his Ph.D. in laser-arc hybrid welding and now currently works. Today, his scope of research covers surface structuring, welding, cladding, cutting, and additive manufacturing. He is the author of over 50 publications, mostly concerning weld phenomena in laser materials processing. Examples in welding are explanations for undercutting, root humping, and arc behavior.

Alexander Kaplan, born in Vienna, Austria (1967), was employed as researcher at Vienna University of Technology from 1989 to 2000. He received his Ph.D. degree (1994) on laser welding modeling. Appointed as the professor and chair of Manufacturing Systems Engineering at Luleå University of Technology, Sweden after a post-doc year at Osaka University (2002). Alexander has particular experience in mathematical modeling and high speed imaging of laser materials processing, particularly for laser welding, laser cutting, and laser additive manufacturing.

Seong Min Hong was born in Gwangju, Korea, in 1988. He got the bachelor degree in the field of international trade and bachelor of economics from Soongsil University, Korea, in 2015. He is a master student under the supervisor of Professor Han-Sur Bang and Professor Hee-Seon Bang in the Department of Welding and Joining Science Engineering at Chosun University.

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Hee-Seon Bang was born in Gwangju, Republic of Korea, in 1971. She received a Ph.D. degree on the development of the laser-arc hybrid welding process in 2011 from the University of Osaka, Japan. Since 2005, she has been the Professor at the Department of Welding and Joining Science at Chosun University, Korea. Hee-Seon Bang has particular experience in welding processes, particularly for FSW, Laser-arc-, and TIG-FSW-hybrid welding.