Development of Heavy Plates for High-Energy Welding of Monopiles for the Construction of Offshore Wind Energy Plants

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ABSTRACT

Welding of large structures, such as monopiles for wind turbines, is very time consuming and thus a key cost factor in the manufacturing process. The use of high heat inputs in submerged arc welding can shorten fabrication time by a multiple, but impairs material properties in the heat affected zone of standard steel grades, in particular toughness. Real multi-layer weldments and physical weld simulations demonstrate the different effectiveness of various kinds of low alloyed heavy plate steels. In-depth microstructure investigations disclose the metallurgical reason for the ability of steels with very low aluminum and silicon contents to resist against high energy welding and confirm the industrial applicability.

Keywords: HAZ, Toughness, TMCP, Offshore, Arctic

INTRODUCTION

For a long time, high energy welding has been a focal point for fabricators to accelerate their manufacturing process for steel constructions. This is particularly true for thick-walled, giant steel constructions, such as the foundations of offshore wind energy plants. The modern urge to increase efficiency has intensified the need for an applicable solution. Due to the large dimensions and high quantities needed for such colossus out of steel, a huge amount of welding time is needed. With the currently common use of submerged arc welding (SAW) the goal is to increase deposition rate and, consequently, the heat input into the base material. Qualification procedures of steel plates for offshore structures according to German Institute for Standardization (DIN) European Committee for Standardization (EN) 10225-1 and respectively American Petroleum Institute (API) Recommended Practice (RP) 2Z describe the requirements for heat inputs up to 5 kJ/mm, but for future applications even higher heat inputs are of interest. This raising thermal load increasingly impairs the microstructure and, therefore, the mechanical properties of the joined plates. Material development aims on the one side to reduce local brittle zones and on the other side to increase the density of effective grain boundaries in the heat affected zone (HAZ).
The focus of this study is the fusion line, since in multi-layer SAW the coarse-grained heat affected zone (CG HAZ) and especially the intercritical reheated coarse grain heat affected zone (ICR CG HAZ) are the most harmful zones.1,2 Figure 01 illustrates the increasing impact of raising heat input on the base material using the measure of HAZ width as an indicator.

Past material developments for high energy welding sought to compensate degradation of properties with high alloying contents3,4 or through complex process control with advanced deoxidation methods5–10. The intention of the latter is to produce small temperature stable non-metallic inclusions of a defined size within the liquid melt to reach a high particle density which can act as nucleus of acicular ferrite. This desirable type of microstructure is known for an increased probability of deflecting a running transgranular cleavage crack due to a high density of high angle grain boundaries, resulting in a large amount of energy dissipation and therefore a good toughness. This is already being used successfully in weld metal of SAW11–14 and has long been considered as the holy grail for HAZ too.

The approach of this study is different and takes up the traditional ideas to avoid the brittle martensite/austenite (MA) microstructure constituents15–18 and to reduce the austenite grain growth due to grain boundary pinning precipitations19,20. One adjustment option involves a low level of elements which exhibit a high solubility in ferrite, like silicon or aluminum. Due to the alloying of these elements the carbon is pushed from ferrite into austenite during the phase transformation and the driving force for cementite precipitation is considerably reduced. This indirectly stabilizes the retained austenite (via the increased carbon content) and enables the formation of MAs.17,21,22 A second option is a proper combination of microalloying system and applied thermomechanical controlled process (TMCP). Niobium is very effective to suppress recrystallisation at low rolling temperatures due to solute drag as well as strain-induced precipitation and thus favorable for achieving plate requirements with lean alloying concepts.23 Titanium also supports the thermomechanical effect, but the advantage lies mainly in the formation of precipitates that are stable up to high temperatures. It is expected that these particles withstand even the short time peak temperature in CG HAZ and are able to limit the temperature-induced growth of austenite grains.24,25 But because of the sub-micrometer size of these precipitations resulting from TMCP, they are not intended to act as nucleation site for acicular ferrite.

EXPERIMENTAL

Several plates of low alloyed steels were investigated in this study. All steels were industrially produced in a continuous casting process, reheated for homogenization in a slab furnace, thermomechanically rolled and subsequently cooled with a high pressure accelerated cooling device. The development of this study aims at a grade S355ML according to EN 10025-4 corresponding to Grade 50 of API 2W. The investigated steels with thicknesses and range of alloying elements are listed in Table 1. It should be noted, that for the steel S355mod the conformity to EN standard is maintained due to the titanium addition, since there is the necessity in the case of very low aluminum contents to add a sufficient amount of other nitrogen binding elements. From this steel 4 heats were produced resulting in an amount of 12 rolled mother plates with slight variation in process parameters. As reference a Dillinger standard steel S355ML with five plates from different heats and as benchmark one plate of S420ML with advanced weldability, proven by fulfilled Arctic 2 conditions26, were used.

| Table 1: Investigated materials with alloying content in wt.% |
|-------------------|---|---|---|---|---|
| thickness | C | Si | Mn | Al | others |
| S355mod | 80 mm | 0.05 | < 0.1 | 1.5 | < 0.01 Nb, Ti |
| S355standard | 62 – 103 mm | 0.06 | 0.4 | 1.3 – 1.5 | 0.03 Nb |
| S420arctic | 65 mm | 0.06 | 0.4 | 1.6 | 0.03 Ni, Cu, Cr, Ti |

For mechanical characterization, tensile tests, Charpy V-notch tests (CVN) and crack tip opening displacement tests (CTOD) were used. Testing was performed in accordance with the standards DIN EN International Organization for Standardization (ISO) 6892-1, DIN EN 10164 and EN ISO 6892-2 for tensile test, DIN EN ISO 148-1 for CVN, and ISO 12135 for CTOD. Mechanical testing was done in transverse direction in the positions sub-surface (S), quarter-thickness (Q) and mid-thickness (M). For toughness testing via CVN or CTOD each data point in the following diagrams represents the mean value of at least 3 single measurements and if the statistics are sufficient, boxplots are created. There, the box is delimited by the upper and lower quartiles and contains the median as horizontal line and a cross as mean value, the antennae extend to the minimum and maximum values.

For physical weld simulation, specimens with a cross section of 10 mm x 10 mm were extracted in transverse specimen orientation from quarter thickness of the plates and subjected to thermal treatment in the thermomechanical simulator Gleeble 3800. For the simulation, the transformation of heat input in cooling time from 800 °C to 500 °C (t8/5) was done in accordance with STAHL-EISEN-Werkstoffblätter (SEW) 088 and transferring from t8/5 into time-temperature profile after Hannerz27.
equation up to a peak temperature of 1350 °C for simulation of CG-region. Furthermore, an additional holding time at peak temperature of 2 s was used in order to tighten the conditions. A second cycle into the two-phase region was performed to simulate ICR CG HAZ, here the $t_{8/5}$ was adjusted according to the remaining temperature for cooling down to 500 °C. After weld simulation the central position of the heat treated zone was notched and a standard CVN test was performed.

Real weldments were done on the one hand by single-layer bead-on-plate and on the other hand by single bevel multi-layer butt welds. For both types of weldments, a tandem SAW machine was used with an OE-SD3 wire and the fully basic agglomerated flux OP 121TT. Welding and mechanical testing were performed in accordance with DIN EN 10225-1 but also with increased heat inputs. Regarding toughness testing, the focus was on specimens with notch location at the fusion line as worst-case scenario.

Simulation of HAZ dimensions was performed using the analytical software MULITPASSE3, which was integrated into the software HAZMET from The Welding Institute. For the calculation of HAZ width only the visual part, without annealing zone, was considered.

Microstructure characterization was performed by light optical and scanning electron microscopy (SEM). This investigation was done on polished longitudinal cross sections, in order to investigate phase constituents, grain structure and particles (precipitations and non-metallic inclusions). Light optical microscope was used for quantification of microstructure constituents, for example MA constituents via Klemm etching and non-metallic inclusions on polished specimens. Cleanliness was evaluated according to DIN EN 10247, supplemented by automated particle analysis in SEM based on backscattered electron (BSE) contrast for particle identification and energy dispersive X-ray spectroscopy (EDS) for chemical characterization in SEM, followed by a classification due to maximum cation and anion concentrations in wt.-%. To cover a statistical representative area, 3.28 mm² were scanned on each specimen. Particle size is specified as circle equivalent diameter. Furthermore, Electron Back Scatter Diffraction (EBSD) analysis was done with two different settings. Because of the fine grain structure on base material an area of 0.16 mm² was measured with a step size of 0.35 μm and on the HAZ an area of 1 mm² with a step size of 0.8 μm. For grain definition a tolerance angle of 5° was used and mean grain size is specified as average value of area fraction distribution function. Kernel average misorientation (KAM) was evaluated with the perimeter of the third neighbors and used for quantification of grain boundary ferrite via peak deconvolution of clearly bimodal distributions of HAZ microstructures. Reconstruction of prior austenite structure was done on HAZ microstructures based on inverse variant selection by ARPGE software and Greninger-Troiano orientation relationship.

Quantification of precipitations generated during TMCP was done by dissolution of steel swarf in hydrochloric acid. Combustion analysis of filter residues and scanning transmission electron microscopy (STEM) of ultra-centrifuged particle suspension were performed. This experimental procedure went along with thermodynamic simulations of the precipitation behavior of microalloying elements with the software package MatCalc, in particular to cover precipitation state during welding.

Lastly, retained austenite was measured via X-ray diffraction (XRD) with a 15 mm aperture and evaluated with Rietveld method.

**RESULTS**

**Mechanical Testing of Base Material**

Tensile test on small round-bar specimens of base material revealed at different positions across plate thickness an increase of yield and tensile strength from mid thickness to the surface, whereas fracture elongation shows in quarter thickness the highest values (Figure 02). In particular tensile strength is for the modified steel slightly lower, but the acceptance relevant testing on rectangular half-thickness specimens is except of one single value all above 470 MPa and mean yield strength is here 390 MPa. Other behavior reveals CVN testing (Figure 03). The modified steel reaches higher values, remarkably is the mid-thickness position at -60 °C since this is the requirement for offshore application (oil & gas) with the addition of applicability in arctic regions (Arctic 2 according to EN 10225-1 Annex F).
Comparable with the fracture elongation, impact energy of quarter thickness is above the energies of sub-surface or mid thickness. Full-thickness CTOD testing confirms the high toughness of S355mod and leads to mean values above 0.3 mm down to -80 °C.

For the additional through-thickness tensile test, the requirement of achieving 80 % of the minimum specified tensile strength is met with an average of 452 MPa, and the quality class Z35 is also significantly exceeded, achieving 58 % reduction of area for the modified steel.

Hot working and cold working properties are illustrated in Figure 04. Hot working is not a common use case of TMCP steel, nevertheless the degradation of strength properties is moderate. Cold working on the other side is a frequently selected option of EN 10225-1 and is not a problem for the modified S355.

Mechanical Testing of Heat Affected Zone
CVN testing of weld simulation specimens demonstrates that, in simulation of CG HAZ as well as ICR CG HAZ, the modified steel exhibits significantly higher toughness values than the standard S355 steel (Figures 05 and 06). The performance even surpasses that of S420arctic steel, especially at heat inputs above 7 kJ/mm. Also, a longer holding time at peak temperature (+2s) does affect S355mod only marginally. The higher values obtained for ICR CG HAZ result from the increased test temperature as an attempt to compensate the smaller area fraction of this region in a real weld. Regarding the peak temperature of the second cycle, there is no clear indication of which temperature is more detrimental.
Real single-layer bead-on-plate weldments result in higher CVN toughness for the modified steel too (Figure 07). In this configuration, only the testing of sub-surface specimens with a notch location comprising 50% HAZ and 50% weld metal (WM) is practicable. At lower heat inputs the difference between S355mod and S355standard is not much pronounced, but it increases with raising heat inputs. The drop in toughness between 2 and 4 kJ/mm is possibly due to the geometrical adaptation of the weld seam with higher deposition rates, which is typical for bead-on-plate weldments. 

Through thickness testing of straight edge fusion line from multi-layer butt welds confirms the superior behavior of the modified S355 steel (Figure 08). Even compared with the S420arctic, the S355mod reaches higher CVN values, even at -80 °C. The tensile test transverse to the welding direction proves material failure within the base material, so weld metal and HAZ do not represent a weak point in this respect. Even full-thickness CTOD at the fusion line, only tested for the 7 kJ/mm, is for the S355 mod steel at a high level (Figure 09). The percentage of CG region is about 25 % in mean for these specimens.
Microstructure Characterization

Despite the very low aluminum and silicon contents in the modified steel, lowest amounts of tramp elements like sulfur ($\leq 0.002$ wt.-%), phosphorus ($\leq 0.01$ wt.-%) and nitrogen ($\leq 0.005$ wt.-%) were achieved. A high cleanliness was measured, in particular no grey (typically sulphides) and no elongated inclusions were detected. A low average area fraction of about 100 $\mu$m$^2$/mm$^2$ of black globular oxides were measured ($K_a$ according to DIN EN 10247) and the light optical determined cleanliness is therefore comparable to that of the S355 standard steel grade. Particles with a size that is an order of magnitude smaller can be quantified by SEM-EDS. Here a distinct feature of the modified steel becomes evident (Figure 10). There are much less particles of the size between 0.5 and 2.0 $\mu$m, but therefore a few more between 2.0 and 5.0 $\mu$m. The ratio of these two particle size classes shows a very significant gap between the two different alloying concepts. Not only the size, but also the classification based on the contained elements is quite different. Whereas for the standard steel most frequently the types AlN with 40 % and CaO with 25 % are classified, the inclusions of the modified S355 are mainly classified as MnS with a percentage of 52 %. But a deeper look especially in the particles of the size class 2.0 to 5.0 $\mu$m reveal a complex chemical structure of this particles (Figure 11). There are a lot of smaller particles like MgAl$_2$O$_4$ or MnS which seems to be embedded in a matrix of CaTiO$_3$. These oxide-based inclusions are not expected to change during heat input from welding, nor has this been observed.

Although S355mod was micro-alloyed with niobium and titanium, no micrometer-sized primary precipitations resulting from a precipitation in the liquid phase during casting were observed at all. But the smaller Ti and Nb precipitations, intended by TMCP, are expected to be significantly below 0.5 $\mu$m and have to be measured in high resolution STEM analysis. With the automatic STEM-EDS measurement of dissolved specimens, several thousands of particles are measured and thus it is suitable for robust characterization of this nanometer-sized precipitations. Figure 12 visualizes the different cation classes respective to their size and relative quantity. Due to the carbon substrate, the classification of anions is not reliable in this method, but the data show a clear predominance of N in Ti and in Al particles as well as C in Nb particles. Although, absolute counts of particles cannot be connected with a particle density, the number of retained particles in the solution is obviously a multiple for the modified steel which is at least an indication that the particle density in the non-dissolved material is higher. In total, the size of particles is mainly between 10 and 50 nm. The measured mean size of all particle types is quite close, but the number fraction of the cation classes differs significantly.

According to combustion analysis, just 27 % of the Ti is bonded in the as-cast condition and not available for TMCP, 48 % precipitates during the process and leads to the measured fine Ti particles. With MatCalc, the solubility temperature of NbC is calculated for S355mod to 1061 °C whereas 1372 °C for TiN, which ensures significantly higher stability of Ti-rich particles in the extreme temperature-affected weld region.
EBSD measurements allow a broad evaluation of quantitative microstructure data. Figure 13 illustrates different outputs from various crystallographic information: image quality (IQ), kernel average misorientation (KAM) and inverse pole figure (IPF) maps. IQ map visualizes grain morphology and internal distortions of crystal lattice in dependance of the pattern quality and is thus roughly related to the classical light optical images of etched specimens. The lattice distortions or misorientations within the grains are represented directly by the KAM and allow an impression of dislocation density. IPF map with a grain boundary overlay illustrates the grains based on crystallographic neighborhood relationships as well as the preferred orientations of crystallographic planes (texture). In particular grain size, dislocation density and phase composition can be extracted from these data. The shown example of S355mod microstructure in quarter thickness reveals a fine-grained structure (regions surrounded by dark lines in IQ and IPF map) of the two phases irregular ferrite with low dislocation density (light in IQ map, blue in KAM map) and a homogenous distribution of bainitic islands with higher internal misorientation (dark in IQ map, green/yellow in KAM map). Rolling texture, induced by TMCP, is not very pronounced in the illustrated quarter thickness position of these thick plates (uniform color distribution in IPF Map). Based on this data, determination of phase fractions at such microstructures is not possible in a reliable way, due to monomodal histograms even of KAM values. However, quantification of grain size and dislocation density, averaged over both phases, can be derived directly and are displayed in Figure 14. For both S355 steels the grain size increases whereas dislocation density decreases from surface to mid thickness. Due to decreasing impact of deformation and cooling in this direction, this corresponds to the expected behavior. Comparing the different steels, the modified S355 exhibit a smaller grain size and a higher dislocation density, resulting from a higher fraction of the bainitic phase.

The appearance of HAZ microstructure is fundamentally different from that of the base material. In real weldments, especially in multi-layer configuration, the localization of representative zones is difficult. In simulated specimens the microstructure is formed by a defined temperature profile which is constant over the hole specimen cross section. Figure 15 illustrates a typical simulated CG HAZ microstructure for S355mod with 22 kJ/mm. Obvious is the dislocation free grain boundary ferrite,
decorating the prior austenite boundaries (light in IQ map, blue in KAM map) and a very dislocation enriched bainite (dark in IQ map, green/yellow in KAM map). The bainite phase appears lath-like (coarse same-colored regions in IPF map with parallel light grey subgrain boundaries) and remarkably not needle-like as anticipated for acicular ferrite. Reconstruction of austenite structure via crystallographic orientation relationship confirms that the visual structure correlates to the prior austenite grains. The mean size increases with heat input, but in both cases the values are significantly lower for the steel S355mod (Figure 16). Surprisingly, there seems to be a tendency for slightly larger austenite grains after a second cycle into the intercritical region. For CG HAZ and ICR CG HAZ the mean austenite grain size of S355mod with 22 kJ/mm is on the same level as the austenite size of S355standard with only 7 kJ/mm.

![Figure 15: EBSD multi-mapping of S355mod with CG HAZ Simulation of 22 kJ/mm](image)

![Figure 16: Mean prior austenite grain size by reconstruction of EBSD-data via ARPGE](image)

Because of the clearly bimodal distribution of KAM values in these microstructures, evaluation of the area fraction of grain boundary ferrite by peak deconvolution can be realized in a simple manner. Grain boundary ferrite is located in the surroundings of prior austenite grains and its area fraction increases with heat input (Figure 17). For ICR CG HAZ it seems that there is in average a slight decrease of grain boundary ferrite. Concerning the materials, S355mod shows a significantly higher area fraction of this phase.

Due to the mostly austenitic structure of MAs, its volume fraction can be quantified in a most robust way via retained austenite measurement by XRD (Figure 18). For standard S355 even with one cycle at 7 kJ/mm there is a noticeable fraction of retained austenite measured. A heating with a second cycle into the two-phase region can further increase this amount. A simulated heat input of 22 kJ/mm on the other side doesn’t show any retained austenite after the second cycle at the latest. The S355mod consistently shows very low values. Also, with EBSD or Klemm etching there were no MAs detected for this steel.

![Figure 17: Phase fraction of grain boundary ferrite determined by peak deconvolution of KAM histogram](image)

![Figure 18: Retained austenite measured by XRD](image)

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DISCUSSION

In this study, with S355mod an uncommon but stable alloying system is presented. One characteristic is the twofold microalloying which can cause problems, but with a stable soft reduction and an adjusted TMCP it has been achieved that no large Ti- or Nb-rich primary precipitations occur in the final product. These coarse particles in the size of a few micrometers are known to be very detrimental to toughness in base material and in particular in HAZ. The second characteristic is the Si-based deoxidation, which has the advantage that the resulting oxides are liquid and can be easily extracted from the steel into the slag. The addition of Ti with strong affinity to oxygen together with inserted Ca for inclusion modification form stable CaTiO3 particles which, as it turns out, are able to capture the smaller impurities and render them harmless. These impurities are composed of trace elements and result from scrap, environment or process reactions.

The usual concerns of insufficient cleanliness and ageing stability due to the absence of an Al alloying can be refuted with the present steel. Light microscopic cleanliness according to the standards is on the same high level as Al deoxidized steels and for smaller inclusions, evaluated with SEM, the particle density is even lower. Mechanical testing at or after treatment at elevated temperatures does not change the properties very much, which proves that nitrogen bonding has sufficiently occurred. Mechanical testing of base material shows on the one side that the strength for S355mod is not far from the lower limit but there is the possibility for improvement by a slight adjusting of TMCP parameters. Toughness testing on the other side demonstrates usability over a wide temperature range. With respect to the weldability, the additional thermal load decreases the toughness with increasing heat input. This not only changes the cooling time as assumed by the formulas in SEW 088, but it is also connected to a longer holding time at peak temperature (compare Hannerz equation). At least from a heat input beyond 7 kJ/mm the superiority of S355mod even over S420arctic becomes apparent. It is important to note, that Gleeble experiments are highly conservative, especially for the execution of two cycles. In the physical simulation, the fracture area consists 100 % of the CG or ICR CG region, which is in reality not the case. Even with testing of straight edge from multi-layer weldment there is just a small amount of these most detrimental HAZ regions within the main fracture plane of toughness specimens. Maximum values of about 30 % are realistic and at least 15 % are required by API RP 2Z. The CVN energies of S355mod in Gleeble simulations match very well with the real multi-layer weldments with 5 and 7 kJ/mm and this confirms the usefulness of physical simulations as a quick forecast. In the end, the real multi-layer weldment and in particular full-thickness CTOD test at fusion line of the straight edge as most severe mechanical test with sufficiently large amount of CG HAZ indicate the applicability of the modified steel at least up to a heat input of 7 kJ/mm. So, there is a very good performance of that steel with regard to base material and HAZ properties since offshore and even arctic requirements are fulfilled. But although all mechanical and technological properties are reached, the modified steel is not in conformity with API Grade 2W since there is no exception for low Al and Si contents.

The reason for the higher toughness of the modified steel can be derived from four essential differences in microstructure:

1. The reduction of MA constituents due to the low Al and Si contents was successful. These well-known local brittle zones were shown to have a very low volume fraction in the case of S355mod. But the difference to the standard steel is limited up to a heat input between 7 and 22 kJ/mm and in particular after performing a two-cycle treatment. In the lower range of heat inputs a longer cooling time means a coarsening of MAs, but at higher heat inputs the low cooling velocities lead to sufficient carbon diffusion and associated with this to a decomposition of retained austenite in less detrimental carbon-rich second phase areas. So, generally MAs seem not to be a major problem for extreme high heat inputs.

2. Controlling the austenite grain growth in HAZ due to small temperature resistant particles is well known. To be effective, the size of particles has to be between 10 and 50 nm and based on the commonly used microalloying elements, the chemical structure should be a titanium compound, NbC is for example not sufficient to resist the high temperatures in CG HAZ. S355mod shows at least in base material a high amount of Ti-rich precipitations with a size clearly below 100 nm and a solubility temperature above 1350 °C according to MatCalc simulations. But direct measurement of this small particles within special regions of HAZ is not possible with the used method of dissolution of steel swarf. The advantage of this technique is the investigation of large volumes to ensure the coverage of a representative volume, but the preparation of very narrow regions of HAZ has to be done by classical target preparation, for example by carbon replica. Nevertheless, the measured austenite grain size has been confirmed to be significantly smaller than the size of the reference steel and due to the chosen area fraction value for comparison, it also ensures a lower heterogeneity of grain size. A minor source of error for the absolute austenite grain size values can be the appearance of grain boundary ferrite, which is difficult to assign to an austenite grain with the used orientation-based reconstruction method. This is possibly an explanation for the measurement of different mean sizes in CG HAZ and ICR CG HAZ.

3. The observation of grain boundary ferrite, also referred to as allotriomorphic proeutectoid ferrite, is typical for the investigated extreme heat inputs. A network of this surrounding phase is assumed to be detrimental to toughness, since it bundles the emerging strains during mechanical testing in a small volume fraction. In this study, quantification of phase fraction with the method of peak deconvolution of KAM distribution works very well for the occurring HAZ microstructure. It is worth mentioning that in the case of the base material, carbon-rich bainitic phases are also visually distinguishable, but there are not two separate peaks in the distribution function allowing a deconvolution. The grain
boundary ferrite is formed during cooling between 800 and 500 °C as a diffusion-controlled process at the beginning of phase transformation until the onset of the displacive transformation inside the austenite grains. The trend for a higher fraction of grain boundary ferrite with increasing heat input is due to the rise in the austenite to ferrite transformation temperature ($A_{c3}$) as well as the longer dwell time within the two-phase region, caused by the lower cooling rate and intensified by its inverse quadratic decrease, described in Hannerz equation. The observed higher phase fraction for S355mod cannot be explained by the chemical differences, since especially the significantly lower Si content leads to a reduction of $A_{c3}$ and this restricts the transformation to grain boundary ferrite. However, the small austenite grain size can have a crucial influence. The increased grain boundary density in the austenite phase favors the nucleation of allotriomorphic ferrite and also accelerates the transformation. It is therefore assumed, that the fraction of grain boundary ferrite would be even higher with increased Si and Al contents. Despite the general appearance of this assumed detrimental microstructure constituent, there is a possible advantage of its grain structure. While especially large, elongated grains of grain boundary ferrite are considered unfavorable, the grains in this study are rather small and equiaxed. This is presumably also an effect of the high nucleation rate due to the small austenite grain size.

4. A major distinguishing feature of the modified steel from standard steels is the appearance of non-metallic inclusions. In particular non-metallic inclusions of the type CaTiO$_3$ arise and the ratio of particles with a diameter of 0.5 to 2.0 µm and particles with diameter of 2.0 to 5.0 µm is below 5. But even if the density of CaTiO$_3$ particles for pervasive nucleation of acicular ferrite is by far too low, they exhibit similar characteristics as the frequently sought particles in the center of acicular ferrite. Like these, the CaTiO$_3$ inclusions remain from the liquid phase and because of their high thermal stability doesn’t change neither due to TMCP nor due to heat input during welding, the size is in the lower micrometer range and there are Mn enriched zones at the outer shell. But there is no Mn depleted zone in the surroundings of the inclusion and no acicular ferrite was detected. Rather, these particles are conspicuous due to their embedment of other particle types. It seems, that the CaTiO$_3$ particles catch other inclusions in the liquid melt and leads with this to a lowering of the total density of micrometer-sized inclusions. The consequence is a larger mean free path between the particles of the size 0.5 to 5 µm and this is related to a higher toughness in the ductile fracture mode. Also, for cleavage fracture a lower number of inclusions can be beneficial, since non-metallic inclusions of that size are frequently initiation point of cleavage transgranular cracks in steels of high light microscopic cleanliness.

To summarize these insights, a linear regression of the identified microstructure parameters with CVN toughness in the heat input range of 7 to 22 kJ/mm is used to reveal the major parameters for microstructure optimization. Based on a least square minimization routine with a joint microstructure fitting for CG HAZ and ICR CG HAZ but separated offset values, there is a good correlation of experimental and calculated CVN energy (Figure 19). It is important to note, that the pre-factors of all microstructure parameters have exhibited negative signs, which means that an increase of each of them leads to a reduction in CVN energy. Based on the observed variation of the individual parameters in this study, the magnitude of the impact on CVN energy can be normalized and quantitatively evaluated (Figure 20). It emerges that a reduction of the total density of non-metallic inclusions between 0.5 and 5.0 µm is by far the most effective parameter. It is followed by a small mean austenite grain size, a low fraction of grain boundary ferrite and lastly a reduced density of MA constituents. This interpretation is valid for an energy range above 5 kJ/mm and for S355 material.

Finally, the small austenite grain size, the low density of micrometer-sized non-metallic inclusions and the absence of MAs are connected with the improved toughness of the modified steel. Contrary to the traditional opinion, the target of resistance against high heat input welding is reached even without the formation of acicular ferrite.

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CONCLUSIONS

A key point for economic fabrication of monopiles for the construction of offshore wind energy plants is the ability of heavy plate steels to resist high heat inputs. In this study a modified steel with very low contents of silicon and aluminum as well as an addition of titanium was investigated. The development target of a steel grade S355ML is far exceeded, since offshore (oil & gas) and even arctic requirements are reached. Contrary to traditional approaches that aimed to achieve high densities of particles in the size range slightly below the classical cleanliness measurements in order to induce intercrystalline nucleation of acicular ferrite, this study shows that also a very low density of non-metallic inclusions can be at least as effective. This indicates that for construction steels, the historical restriction to only aluminum for deoxidation and nitrogen bonding is no longer justified. With the right application of TMCP technology, there is a high potential of this kind of steel for high-efficient joining especially in offshore wind industry. In addition, the improved toughness up to extreme heat inputs is also suitable for advanced welding processes like electrogas or electroslag welding. Consequently, there is a clear recommendation for API to reconsider their minimum request for aluminum and silicon contents. The proposed steel is of lean chemistry, suitable for mass production and can contribute to achieving the climate targets in Europe and possibly beyond.

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