RESIDUAL STRESS ANALYSIS IN WELDED JOINTS
BY NEUTRON DIFFRACTION AND COMPUTER MODELING

GIZO BOKUCHAVA¹,*, PETER PETROV², GANCHO GENCHEV³, VESSELIN MICHAILOV³, NIKOLAY DOYNOV³, RALF OSSENBRINK³

¹ Frank Laboratory of Neutron Physics, Joint Institute for Nuclear Research, Joliot-Curie Str. 6, 141980 Dubna, Russia,
² Institute of Electronics of Bulgarian Academy of Sciences, 72 Tzarigradsko Chaussee, 1784 Sofia, Bulgaria,
³ Department of Joining and Welding Technology, Brandenburg University of Technology, Platz der Deutschen Einheit 1, 03046 Cottbus, Germany,
E-mails: pitiv@ie.bas.bg; gancho.genchev@b-tu.de; michailov@b-tu.de; doynov@b-tu.de; ralf.ossenbrink@b-tu.de
*Corresponding author, E-mail: gizo@nf.jinr.ru

Received December 4, 2017

Abstract. Numerical models for thermo-mechanical simulation of the welding residual stresses in laser beam and multi-pass arc welded joints are developed. The models describe the resulting microstructure as a function of the peak temperature, austenization time, cooling time and take into account the microstructure formed after single or multiple thermal cycles. The results of simulations are validated by neutron diffraction experiments performed on FSD diffractometer at the IBR-2 pulsed reactor in FLNP JINR (Dubna, Russia). The elaborated numerical models are in good agreement with obtained neutron diffraction data, which enable to study the influence of different conditions and process parameters on the distribution of residual welding stresses.

Key words: neutron diffraction, residual stress, welding, computer modeling.

1. INTRODUCTION

Complex structures often are assembled using various welding techniques. As a result of the heat exposure, the local domains of the heat-affected zone (HAZ) are subjected to cumulative thermo-mechanical influences. This affects in a complex way the microstructure transformations and thereby the local thermo-mechanical properties and the forming of residual welding stresses which can significantly reduce the quality and reliability of the welded components. Therefore, the accurate prediction of the residual stress distribution is of a major importance for the precise design of the welded structure in order to improve the quality and reliability of the welds.

The numerical simulation based on theory of welding heat effects [1] allows to obtain relatively quick results of the final state considering the thermo-mechanical history as well. However, the accuracy of the results strongly depends on the mathematical modeling complexity of the welding process as well as on the
applied models and input data. Numerical models used in computational welding mechanics exist for a wide range of problems such as welding distortions, residual stresses, HAZ microstructure composition. The models for the microstructure transformation of steels are based on the cooling time concept. However, existing models are developed primarily based on correlations of single temperature cycles, and, respectively, on single-pass welding. For this reason, they often demonstrate significant discrepancy between calculation results and measurements. Moreover, in contrast to the maximum temperature austenization cooling time concept (STAAZ model) [2], they are not able to predict the changes of the microstructure and mechanical properties over the sub-domains of the HAZ according to different peak temperatures during the repeated heat treatment in multi-pass welds.

After first welding pass of structural low-alloyed steels, the microstructure of the base material is subjected to the thermal influences and as a result a different perlite, ferrite, bainite, martensite respectively mixed microstructure in the HAZ can be formed. This local microstructure state is initial for the next welding pass and because of the repeated heating a different microstructure with changed properties is formed as detected in [3]. Thus, after the next welding pass, the resulting microstructure is subjected to repeated thermal influences. Therefore, the final microstructure in the HAZ of multi-pass welded joint depends on the full thermal history [4, 5]. The influence of the local interpass microstructure on the resulting thermo-mechanical properties, particularly after re-austenization, and its impact on the residual stress distribution is not enough investigated up to now. Since the multi-pass arc welding leads to 3D stress state with inhomogeneous stress distribution and gradients in the thickness, several locations in the welded joint should be considered for the stress evaluation.

The aim of this work is to develop a validated simulation models for the investigation of residual welding stress distribution in single- and multi-pass welds. The simulation is focused on the consideration of the local microstructure properties changes due to the multiple welding thermal cycles. Therefore, the former suggested maximum temperature austenization cooling time approach (STAAZ model) is extended for repeated temperature cycles (M-STAAZ model). The model describes the resulting thermo-mechanical properties as a function of the local temperature welding cycles, characterized by the three relevant model parameters – maximum temperature $T_{\text{max}}$, austenization time $t_A$ and cooling time $t_{8/5}$. The extended multiple maximum temperature austenization cooling time approach (M-STAAZ) takes into account the resulting microstructure properties formed after the preliminary welding thermal cycle as initial state for the next welding thermal cycle. Physical Gleeble® simulation of the HAZ microstructure is performed and experimental material data are obtained for the model calibration.

For the simulation model validation, a high precision experimental data are required. The measurement of residual stress distribution is very difficult due to the complex 3D stress state in the welds. Therefore, experimental methods for
determination of residual stresses such as the X-ray diffraction or hole drilling method are not suitable for this purpose, since they provide results on the surface of the specimen. For these reasons, the neutron diffraction method is chosen, which allows to investigate residual stress distribution at the depths of several centimeters in the bulk of the material due to high penetration power of the neutrons.

2. MATERIALS AND METHODS

2.1. SAMPLES AND WELDING PARAMETERS

Two samples welded with different welding techniques were investigated in this research work. The first one is thin C45 low-alloyed steel plate with dimensions of 100 × 100 × 2.5 mm³ (sample LBW). The sample was welded using a solid-state laser with heat input per unit length of 200 J/mm (Fig. 1a), the focus diameter of 1.44 mm and the beam focus position at the bottom side of the specimen (z = 0). No shielding gas and no filler material were used for welding experiments. With this parameters were performed. The sample was mounted free on three points and welded by laser beam using a 6-axis robotic arm. In order to verify the temperature field simulation, thermocouples were installed to measure the temperature history on the top surface and on the bottom side at different locations near the weld. The neutron measurements of residual stresses were performed in the middle part of the plate across weld seam line.

The second studied sample is a multi-pass butt welded joint of the low alloyed steel S355J2 + N (Fig. 1b) with dimensions of 500 mm length, 150 mm (× 2) width and 20 mm thickness (sample GMAW/SAW). For the welding experiments, Gas Metal Arc Welding (GMAW) method was used for the first welding pass and Submerged Arc Welding (SAW) method is used for the second and third welding passes. Parameters for GMAW welding of the first weld pass were the following: current \( I = 280 \, \text{A} \); voltage \( U = 28 \, \text{V} \); welding speed \( V = 65 \, \text{cm/min} \); filler material G4Si1 (1.5130), \( \varnothing 1.2 \, \text{mm} \) / shielding gas M21 (80% CO₂ + 20% Ar). For SAW welding of the second and third weld passes, the following parameters were used: current \( I = 400 \, \text{A} \); voltage \( U = 29 \, \text{V} \); welding speed \( V = 50 \, \text{cm/min} \); filler material OE – S2 (EN 756 – S2), \( \varnothing 3 \, \text{mm} \) / welding flux OP 181 (EN 760). The yield strength \( R_{p0.2} = 385 \, \text{MPa} \) and Vickers hardness of 165 HV are experimentally determined for the studied sample’s material. The initial sample temperature as well the interpass temperature were always equal to room temperature. With this configuration welding experiments were performed and temperature cycles measurements by means of thermocouples were carried out. For neutron diffraction experiments, a specimen was cut from the entire multi-pass welded joint (Fig. 1c). The neutron measurements were performed at the middle of the sample’s width at the depth of 7.5 mm from sample’s bottom surface.
2.2. RESIDUAL STRESS MEASUREMENTS BY NEUTRON DIFFRACTION

Diffraction of thermal neutrons is one of the most informative tools in materials science and engineering due to high penetration power of the neutrons, non-destructive character of the method and possibility of precise structural and microstructural analysis of modern multiphase materials [6, 7].

Internal stresses, e.g. residual welding stresses, existing in a material cause corresponding lattice strains, which, in turn, results in shifts of Bragg peaks in the diffraction spectrum. This yields direct information on changes in interplanar spacing in a gauge volume, which can be easily transformed into data on internal stresses, using known elastic constants (Young’s modulus) of a material. The principle of the determination of the lattice strain is based on the Bragg’s law

\[ 2d_{hkl} \cdot \sin \theta = \lambda, \]

where \( \lambda \) is the neutron wavelength, \( d_{hkl} \) is the interplanar spacing, and \( \theta \) is the Bragg angle. In this case, the lattice strain is determined as

\[ \varepsilon_{hkl} = (d_{hkl} - d_{hkl}^0) / d_{hkl}^0 = -\Delta \theta / \cot \theta = \Delta t / t, \]

where \( d_{hkl} \) is the measured interplanar spacing and \( d_{hkl}^0 \) is the same interplanar spacing in a stress-free material, and \( t \) is the neutron time of flight.

The components of the residual stress tensor can be determined from the measured residual strain according to the Hooke’s law:

\[ \sigma_{ij} = \frac{E}{1 + \nu} \cdot (\varepsilon_{ij} + \frac{\nu}{1-2\nu} \cdot (\varepsilon_x + \varepsilon_y + \varepsilon_z)), \]

where

- \( \sigma_{ij} \) is the stress in direction \( ij \),
- \( E \) is the Young’s modulus,
- \( \nu \) is the Poisson’s ratio,
- \( \varepsilon_{ij} \) is the strain in direction \( ij \),
- \( \varepsilon_x, \varepsilon_y, \varepsilon_z \) are the strains in the axial directions.

Fig. 1 – a) Thin steel plate LBW sample with laser beam welded joint. b) Multi-pass butt welded joint. c) GMAW/SAW specimen for the neutron diffraction study prepared by cutting from the multi-pass welded joint.
where \( \sigma_{ii} = X, Y, Z, \) and \( \varepsilon_{ii} \) are components of the stress and strain tensors, respectively, \( E \) is Young’s modulus and \( \nu \) is Poisson’s ratio.

The essence of the diffraction method for studying stresses is rather simple and, in conventional experimental scheme, consists of incident and scattered neutron beam shaping using diaphragms and/or radial collimators and definition of a small scattering volume (gauge volume) in the bulk of the specimen (Fig. 2) [8]. Usually the incident beam is formed using cadmium or boron nitride diaphragms with characteristic sizes of 1–2 mm. To define a gauge volume of optimum shape in the studied specimen, at the scattered beam radial collimators with a lot of vertical slits formed by Mylar films with gadolinium oxide coating are often used. A radial collimator is placed at a quite large fixed distance (~150 ÷ 450 mm) from the specimen and provides good spatial resolution of the level of 1–2 mm along the incident neutron beam direction.

![Diagram](image)

Fig. 2 – a) Scheme of the experiment for residual stress investigation in a bulk object. Incident and scattered neutron beams are restricted by diaphragm and radial collimators shaping a gauge volume within the specimen. b) Sample place at FSD diffractometer. Goniometer, the diaphragm at the incident beam and two radial collimators at the scattered beams are visible.

The lattice strain is measured in the direction parallel to the neutron scattering vector \( Q \). The sample region under study is scanned using the gauge volume by moving the sample in the required directions. In this case, relative shifts of diffraction peaks from the positions defined by unit cell parameters of an unstrained material are measured.

### 3. RESULTS

3.1. MATHEMATICAL MODELING OF LASER BEAM WELDING PROCESS

Because of the symmetry of the problem, a half model of the specimen is used. For correct description of the thermo-mechanical behavior of the metal in the
HAZ a quite fine mesh is necessary. The mesh in the HAZ has an element size of 0.37 mm in X- and Y-direction and 0.625 mm in Z-direction. The first step of the thermo-mechanical numerical simulation of welding is the temperature field calculation. It includes the mathematical modeling of the heat input of the welding process with cylindrical Gaussian heat source model and accurate thermo-physical material properties as function of the temperature.

The next step is the modeling of the thermo-mechanical behavior during welding. In particular, microstructure transformations are very important for welding of steels since they can affect strongly the residual stress level and its distribution. The methodical procedure for the thermo-mechanical FE-simulations is based on the maximum temperature austenization cooling time approach. The STAAZ model describes the resulting material properties as a function of the local temperature welding cycles, characterized by the three relevant model parameters – maximum temperature $T_{\text{max}}$, austenization time $t_\text{a}$ and cooling time $t_{8/5}$. The model is qualified for beam welding with high heating rates which the model considers by short $t_\text{a}$ times. For the correct model calibration, material properties obtained by beam welding conditions are required. For this purpose thermo-mechanical material properties considering the high temperature gradients and microstructure transformations at relevant maximum temperatures during laser beam welding are performed. The HAZ microstructure during rapid welding temperature cycles is physically simulated using Gleeble 3500. Experimentally obtained thermal strain curves during heating up to maximum temperatures $T_{\text{max}} = 1000$ °C as well as $T_{\text{max}} = 1300$ °C and cooling down with $t_{8/5}$ of $\approx 1$ s are used for the calculation of the thermal expansion coefficient for the base and HAZ materials (Fig. 3a).

![Graph](image-url)

Fig. 3 – a) Thermal expansion coefficient. b) Yield strength and Young’s modulus vs. temperature.

In further experiments tensile tests of the physically simulated microstructures are performed. The yield strength at room and higher temperatures is used in order to describe the mechanical properties of the HAZ. Fig. 3b shows the temperature dependent yield strength $R_{\text{p0.2}}$ for the base material during heating and for the HAZ materials during cooling. For the elasticity modulus, literature data are used and the
same temperature dependences are defined for the material data during heating and cooling. With the obtained material data and using the STAAZ model and the FE software ANSYS® thermo-mechanical simulations are performed and the distribution of the residual welding stresses after welding is determined at room temperature (Fig. 4).

![Fig. 4 – FEM calculation results: residual stress distribution map after laser beam welding for longitudinal X (a) and transverse Y (b) components.](image)

The residual stress distribution in the welded joint was investigated by neutron TOF diffraction method. The experiments were performed on Fourier stress diffractometer FSD at the IBR-2 fast pulsed reactor in FLNP JINR (Dubna, Russia) [9]. The FSD diffractometer was specially designed for residual stress studies in bulk industrial components and new advanced materials [10]. On FSD, a special purpose correlation technique is used to achieve high resolution level of the diffractometer – a combination of the fast Fourier chopper for the primary neutron beam intensity modulation and the reverse TOF method for data acquisition [11]. This allows to obtain high-resolution neutron diffraction spectra $\Delta d/d \approx 2 \cdot 10^{-3}$ for backscattering detector ($2\theta = 140^\circ$) and $\Delta d/d \approx 4 \cdot 10^{-3}$ for $\pm 90^\circ$ – detectors at $d = 2$ Å with short flight distance (~6.6 m) between Fourier chopper and detectors.

During the experiment on the FSD neutron diffractometer a small scattering volume (gauge volume) of the size of $2 \times 2 \times 10$ mm$^3$ was defined using radial collimator in front of the $90^\circ$ detector. Orienting a specimen in a certain way with respect to the direction of the neutron scattering vector the main strain components were measured during the scan across weld regions. All main diffraction peaks of the spectra were indexed in the frame of the BCC structure (sp. gr. $Im3m$) with lattice constant $a \approx 2.87$ Å. The measured diffraction spectra were processed using full profile analysis based on the Rietveld method (Fig. 5). The average residual strain was determined as $\varepsilon = (a - a_0) / a_0$, where $a$ is the lattice parameter for studied
specimen with residual stresses and $a_0$ is the lattice parameter for stress-free reference material. From the obtained strain values the residual stress components were determined according to Eq. (4).

![Typical neutron diffraction pattern from LBW plate sample, measured on the FSD diffractometer by 90° detector with radial collimator. The experimental points, the profile curve calculated by the Rietveld method and difference curve are shown. b) 3D intensity map near (110) reflection: peak broadening effect at weld seam position (X = 0 mm) is clearly seen.

Fig. 5 – a) Typical neutron diffraction pattern from LBW plate sample, measured on the FSD diffractometer by 90° detector with radial collimator. The experimental points, the profile curve calculated by the Rietveld method and difference curve are shown. b) 3D intensity map near (110) reflection: peak broadening effect at weld seam position (X = 0 mm) is clearly seen.

Fig. 6 demonstrates the results of the measurements of welding residual stresses by means of neutron diffraction and comparison with the numerical simulation results. The results represent the distribution of the stresses across the welding direction along a path on the middle of the length ($Y = 50$ mm) and in the middle plane ($Z = 1.25$ mm) of the specimen. The results show a good agreement of the numerical calculated distribution of longitudinal stresses with the measurements. The longitudinal residual stresses have a peak at distance of approx. 3 mm from the weld axis. This area corresponds to the end of the HAZ of the welded joint respectively the transition to the base material. The peak is well represented by the simulation as well as by the measurement. In the area of $0 < X < 3$ mm, corresponding to the HAZ, the longitudinal stresses decrease significant due to the martensite microstructure formed in the HAZ including the weld seam. The obtained distribution of the longitudinal stresses is in good agreement with literature data for similar steels [12].

The results for the transverse residual stresses indicate qualitative good agreement between measurements and simulation. The calculated distribution of the transverse residual stresses far from the HAZ as well as in the HAZ is verified by the measurements. In the weld metal area the measurement shows higher transverse stresses in comparison with the simulation. The difference between measurements and simulations particularly in the weld seam, in both longitudinal
Residual stress analysis in welded joints

and transverse stress distribution can be explained with the different microstructure (presence of residual microstrain defined by high dislocation density, average grain size changes, etc.) of the weld metal. However, the weld metal microstructure variation is not taken into consideration in the mathematical modelling and numerical welding simulation.

Fig. 6 – Comparison of residual strain (a) and stress (b) tensor components measured by neutron diffraction with FEM calculations. Symbols correspond to experimentally measured strain and stress values while solid lines are results of FEM calculations.

Another important possibility for microstructure investigation gives an analysis of the shape of a diffraction peak (in the simplest case, its width) which can provide data on crystal lattice microstrains inside individual grains and crystallite sizes [13]. TOF diffractometers at pulsed sources have a good potential for the modern materials microstructure characterization due to simplicity of the functional relationship between the instrument resolution $R(Q)$ and the momentum transfer $Q$, which is almost independent on $Q$ within a fairly wide range. As a rule, TOF instruments exhibit a rather wide range of interplanar spacing and, consequently, possess a large number of simultaneously observed diffraction peaks with the almost similar contribution of the resolution function to their widths. This enables one to estimate lattice microstrain and the size of coherently scattering domains (crystallites) from diffraction peak widths in a rather simple way [14, 15].

Significant diffraction peak broadening effects are often observed in welded joint and HAZ regions observed due to the change in the material microstructure during weld process. Usually these effects are caused by crystallite size changes due to material recrystallization after welding and by the increase in the residual lattice microstrain, which directly characterizes the dislocation density in a material. From the broadening of diffraction peak widths, the level of residual microstrain was evaluated for studied sample (Fig. 7). It was found that the microstrain distribution exhibits sharp maximum at weld seam position ($X = 0$ mm)
with maximal level of $\sim 4.8 \cdot 10^{-3}$. At the same time a sharp gradient in microstrain (down to $\sim 1 \cdot 10^{-3}$) is observed while moving from weld center to the HAZ region’s edge. The estimated dislocation density varies from $\sim 2 \cdot 10^{14}$ m$^{-2}$ for the base material to the maximal value of $\sim 5.4 \cdot 10^{15}$ m$^{-2}$ at the weld seam center.

![Graphs of microstrain and dislocation density](image)

**Fig. 7** – Lattice microstrain (a) and dislocation density (b) estimated from peak broadening.

### 3.1. ADVANCED MODEL FOR MULTI-PASS BUTT WELDING SIMULATION

The methodical approach for this study is based on experimental and numerical steps. It includes the mathematical modeling of the heat input of the welding process with eligible heat source models and accurate thermo-physical material properties as function of the temperature. The first step of the thermo-mechanical numerical simulation of welding is the temperature field calculation performed by means of transient *Finite Element Analysis* (FEA). For the modeling of the heat input during the welding processes volume heat source with Gaussian heat distribution is applied. The thermal FEA are validated using micrographs of the welded joints and temperature cycles measurements with thermocouples obtained in the welding experiments. In Fig. 8 the calculated maximum temperatures on the bottom and top sides of the GMAW/SAW specimen and comparison with the measurements by thermocouples for each welded pass are shown. The calculated maximum temperatures show agreement with the measurements and deviations $< 15 \%$. With the validated simulation model a further post-processing is performed and the parameters of the M-STAAZ model – maximum temperatures $T_{\text{max}}^i$, and cooling times $t_{8/5}^i$, *i.e.* from 800°C till 500°C for each welding pass ($i = 1, 2, 3$) are identified as follow: 1) cooling times of $2 \leq t_{8/5}^1 \leq 4$ s for the 1$^{\text{st}}$ GMAW welded pass; 2) cooling times of $6 \leq t_{8/5}^2 = t_{8/5}^3 \leq 12$ s for the 2$^{\text{nd}}$ and the 3$^{\text{rd}}$ SAW welded passes.
The austenization time is related to the maximum temperature and its variation is not considered explicit. The microstructure properties formed after the preliminary thermal cycle \((i-1)\) is considered by the extended approach through the parameters \(T_{max}^{i-1}\) and \(t_{8/5}^{i-1}\) as initial state for the current thermal cycle \((i)\). Thereby the interpass microstructure properties are taken into account. The correlation between the model parameters \(T_{max}\) and \(t_{8/5}\) and the thermo-mechanical properties of the microstructure is determined by experimental data. For the model calibration a database with material data sets for different HAZ microstructure states, reproduced by means of physical simulation is used. The physical simulation of the HAZ microstructure is conducted with Gleeble® 3500. As input parameters for the Gleeble® simulations the experimentally identified model parameters in the HAZ of the multi-pass welds are used. The base material and two further interpass microstructure states representative for the coarse grain HAZ are subjected to repeated temperature cycles: 1) interpass state I: \(T_{max} = 1250^\circ C\) and \(t_{8/5} = 2\) s; 2) interpass state II: \(T_{max} = 1250^\circ C\) and \(t_{8/5} = 6\) s.

Dilatometer and tensile tests are conducted and the thermal strain respectively microstructure transformation temperatures and the yield strength \((R_{p0.2})\) are
determined for the characteristic microstructures (interpass state I and II). Fig. 9a shows the measured evolution of thermal strain during the welding thermal cycle with the determined temperatures of $\gamma \rightarrow \alpha$ phase transformation (beginning at Ar3 and ending at Ar1). Since the phase transformations occur at high temperatures, the microstructure yield strength has decisive significance for the residual stress formation. The yield strength curves in Fig. 9b, c are based on results from tensile tests during heating and cooling. The yield strength development between the test points and the transformation temperatures Ar3 – Ar1 is assumed linear. As a result, the yield strength evolution is obtained during single heating and cooling of the base material as well as during repeated heating and cooling of the two representative interpass microstructure states.

In the next step, thermo-mechanical simulations of the multi-pass welded joint are performed using the obtained material properties data for the model calibration. First, the changing of the local microstructure properties ($R_{p0.2}$ and temperatures of phase transformations, i.e. thermal strain) after each welding pass are calculated by the M-STAAZ model using numerical interpolation. The derived microstructure properties are subsequently applied for the calculation of the residual welding stresses in the multi-pass joint by elastic-plastic FE analysis using bilinear kinematic hardening model with temperature independent tangent modulus of 1000 MPa. Additionally, the relaxation of the stress state due to specimen cutting process is simulated using the common element deactivation technique.

The repeated heating and cooling has a complex influence on the local properties changes and on the accumulation of the plastic strains and thus on the residual stress formation. To identify the influence of the local properties, particularly the yield strength distribution, two simulation approaches are applied. They differ only in the consideration of the local interpass microstructure properties as follow: 1) M-STAAZ takes into account the multiple heating effects on the interpass microstructure properties and their influence on the resulting yield strength distribution (see Fig. 9c); 2) STAAZ uses thermo-mechanical properties for single heating of the base material and single pass welding (see Fig. 9b).

The simulations with the two different approaches show similar results (Fig. 10a) in the areas far from HAZ and weld joint center. In the regions outside the HAZ there is no influence of the multiple thermal cycles on the microstructure properties as confirmed by the yield stress calculations. In contrast to that, the simulations indicate differences for the calculated stresses at the locations of the welded passes particularly in the root pass and it’s HAZ. The root pass is welded with GMAW and consequently characterized by short cooling times of $2s \leq t_{6.5} \leq 4s$. The subarea of the first pass, which is subjected to temperature cycles in a range of $T_{\text{max}} = 1000^\circ\text{C}$ and cooling rates of $t_{6.5} = 6s$ by the second welded pass, has higher yield strength $R_{p0.2}$ when considering the material properties change according to the M-STAAZ model (see Fig. 9c). In contrast, for the simulation with approach 2 material properties for single temperature cycle according to Fig. 9b, are used and the $R_{p0.2}$ increase is not considered. Applying the derived microstructure
properties for the residual stress simulations leads to dissimilar distribution of the calculated residual stresses with two different approaches.

Fig. 9 – Measured thermal strain and temperatures of microstructure transformations (a). Evolution of yield strength during heating and cooling of the base S355J2+N steel material (b) and two interpass microstructures (c).
During the experiment on the FSD diffractometer a gauge volume of the size of $5 \times 2 \times 20$ mm$^3$ for $Y$- and $Z$-components and of the size of $5 \times 2 \times 5$ mm$^3$ for $X$-component was defined in the depth of the studied specimen using radial collimator in front of the 90° detector. In the investigated specimen, the residual stress tensor components distributions over scan coordinate $X$ alternate in sign and vary within wide limits (Fig. 10b). The maximum of the residual stress distribution agree rather well with the position of the weld seam. As would be expected, the residual stresses decrease sharply with distance from the welded-seam region.

![Fig. 10 – a) Comparison of simulation results within STAAZ and M-STAAZ models: yield stress change $R_{p0.2}$ (a), distribution of longitudinal (b) and transverse (c) residual stresses. b) Residual stress distribution measured by neutron diffraction.](image)

In Fig. 11 and Fig. 12 the residual stresses calculations with the two approaches are compared to the measurements along the path “S”. These results are determined on the specimen, which was cut from the entire multi-pass welded joint. The numerical results with the M-STAAZ model for both – longitudinal and transverse residual stress distribution are in agreement with the measurements. The experimentally obtained width of the tensile residual stresses and the peaks of the stresses are realistically calculated by the simulations using the M-STAAZ model (approach 1). However, the calculations with STAAZ (approach 2) show higher deviations in comparison to the measurements for both the longitudinal (Fig. 11) and transverse (Fig. 12) residual stresses, particularly in the HAZ and the weld seam. The measurements at location more than ±10 mm from the weld center show deviation compared to the simulations. These measuring points are located in the area of recrystallization, where the HAZ microstructure is changing to the ferritic-pearlitic microstructure of the base material.

The difference in numerical results between M-STAAZ and STAAZ models can be explained with the influence of the local thermo-mechanical properties in the multi-pass weld on the forming of the welding residual stresses. Due to the thermal welding history the interpass microstructure is subjected to reaustenization and tempering effects and the yield strength $R_{p0.2}$ change respectively after each welding
pass. In the simulations with material properties for single temperature cycle according to the STAAZ approach the $R_{p0.2}$ change is not considered. In contrast, the M-STAAZ model predict the higher $R_{p0.2}$ after the reaustenization during the next welding pass, as obtained in the physical simulation. As a result of the considered yield strength increment and its temperature dependence the calculated longitudinal as well as transverse residual stresses increase significantly and the agreement with the measurements improves.

![Longitudinal stress distribution](image1.png)

![Transverse stress distribution](image2.png)

**Fig. 11** – Longitudinal stress $\sigma_X$ distribution calculated within M-STAAZ approach (a) and (b) comparison of the simulation results with measured data.

**Fig. 12** – Transverse stress $\sigma_Y$ distribution calculated within M-STAAZ approach (a) and (b) comparison of the simulation results with measured data.

4. CONCLUSIONS

In this work, the residual welding stresses are investigated experimentally and numerically. For the mathematical modelling of the thermo-mechanical behavior
during welding the STAAZ and M-STAAZ approaches and experimentally obtained thermal strain curves and temperature dependent yield strength are used. The FEM simulations results indicate good agreement with neutron diffraction measurements. The provided numerical models enable to study the influence of different conditions and process parameters on the development of microstructure properties, residual stresses and distortions during laser beam and multi-pass arc welding. The suggested simulation approach considering the local microstructure properties establishes the basis for numerical fatigue analysis of welded components under consideration of the local material properties, residual stresses and manufacturing induced damage due to welding.

Acknowledgements. The work was supported within the Romania-JINR Programme 2017. P. Petrov acknowledges support from Bulgarian Nuclear Regulatory Agency. G.D. Bokuchava acknowledges support from Russian Foundation for Basic Research (RFBR) within project No. 15-08-06418_a.

REFERENCES