

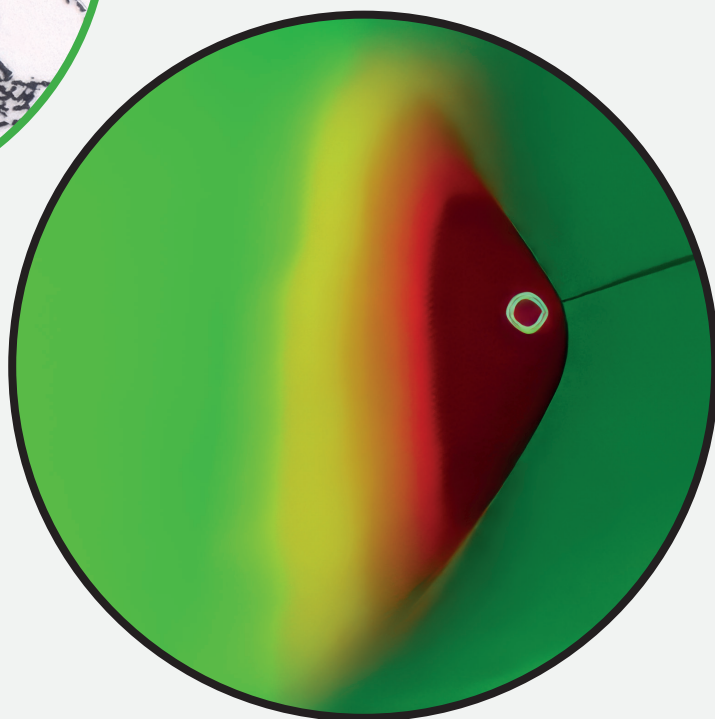
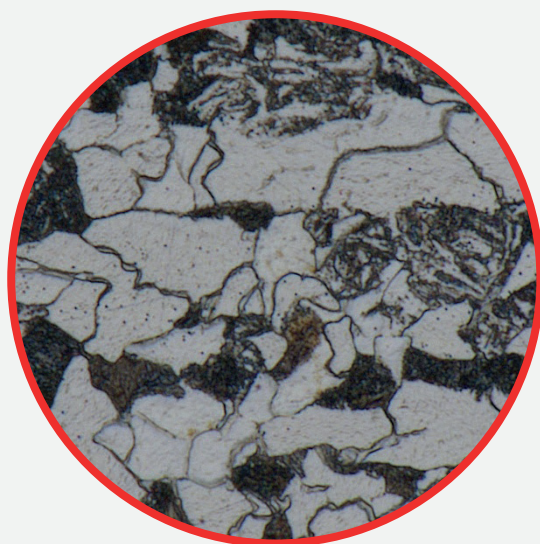
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Wydawnictwa AGH (AGH University Press)

al. A. Mickiewicza 30, 30-059 Kraków

tel. 12 617 32 28, 12 638 40 38

e-mail: redakcja@wydawnictwoagh.pl

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A Methodology for Evaluating Dimensional Accuracy in Reverse Engineering Workflows

Kristína Kovalčíková^a , Marek Šolc^{a*} , Dominik Dubec^a , Tomasz Małyś^b 

^a Technical University of Košice, Faculty of Materials, Metallurgy and Recycling, 1/9 Letná St., 042 00, Košice, Slovakia

^b Silesian University of Technology, Faculty of Materials Engineering, 8 Krasinskiego St., 40-019 Katowice, Poland

*e-mail: marek.solc@tuke.sk

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Abstract

Reverse engineering is widely used for the digital reconstruction of physical objects; however, the dimensional accuracy of the resulting models depends on a sequence of interconnected processes, including manufacturing, 3D scanning, data processing, and model reconstruction. Most existing studies evaluate individual devices or technologies separately, while the cumulative influence of the entire workflow is rarely addressed in a systematic manner. This paper proposes a structured methodology for evaluating the dimensional accuracy of reverse engineering systems as integrated processes. The proposed framework defines a repeatable workflow consisting of reference artifact design, additive manufacturing, object digitisation, mesh processing, CAD reconstruction, and dimensional evaluation. The methodology provides a general framework that can be applied to different reverse engineering technologies and measurement systems. Its objective is to support consistent accuracy evaluation, improve the comparability of results across studies, and contribute to the development of standardised procedures for assessing reverse engineering workflows.

Keywords:

3D scanning, accuracy evaluation, measurement, mesh processing, engineering

1. INTRODUCTION

Reverse engineering based on 3D scanning has become a widely used approach in product development, quality control, and digital reconstruction of physical objects. Modern structured-light and laser scanners allow rapid acquisition of complex geometries, which can subsequently be transformed into editable CAD models. Despite the technological maturity of individual devices, the overall accuracy of reverse engineering (RE) processes remains difficult to evaluate in a consistent and repeatable way [1–4].

Most manufacturers specify the accuracy of their scanners using proprietary procedures or reference standards focused solely on the scanning device. International standards such as VDI/VDE 2634 define procedures for testing optical 3D measuring systems, but these standards address only the performance of the scanner itself under controlled laboratory conditions [5]. They do not consider the complete reverse engineering workflow, which typically includes multiple stages: data acquisition, mesh processing, conversion into CAD geometry, and often subsequent reproduction using additive manufacturing [6]. As a result, the declared scanner accuracy does not necessarily reflect the dimensional fidelity of the final reconstructed model.

Several researchers have pointed out that the accuracy of 3D scanning is strongly influenced not only by hardware

parameters but also by external and procedural factors. Kersten et al. demonstrated that operator experience, calibration quality, object positioning, and lighting conditions significantly affect the resulting measurement accuracy [1, 7–9]. These influences introduce variability that is not captured by standard device specifications.

Similarly, Campanelli et al. emphasised that surface properties, scanning strategy, and data processing steps can cause deviations between the real object and its digital representation. Their work highlights that the error is cumulative and propagates throughout the workflow rather than being isolated to the scanning stage alone [6].

Medřický et al. further analysed the influence of environmental conditions and post-processing software on the final mesh quality. Their findings indicate that smoothing, simplification, and filtering algorithms can significantly alter the geometry of the scanned object, often improving the visual quality at the expense of dimensional accuracy [5]. This introduces an important challenge for reverse engineering applications where dimensional fidelity is critical.

As previously mentioned, reverse engineering often includes more stages, where each of these stages introduces its own type of error, and their combination determines the overall dimensional accuracy of the process [10–13]. However, there is currently no standardised approach

describing how such a system should be tested, what type of test object should be used, how measurements should be performed, or how the results should be statistically evaluated [1].

For evaluation purposes, some authors have used simple geometric artifacts (spheres, cubes, cylinders) to test scanner accuracy. While suitable for device calibration, such objects are insufficient for assessing reverse engineering workflows, where complex geometrical features and transitions between simple and detailed surfaces play a crucial role [14–16]. Therefore, the design of a dedicated test artifact containing a combination of simple and fine geometrical elements becomes essential for meaningful evaluation.

Moreover, a traditional dimensional comparison between the original and reconstructed model is often limited to a small number of linear measurements [17]. This approach does not fully describe the geometric fidelity of the surface [18, 19]. The use of statistical indicators, such as the root mean square error obtained from best-fit alignment in inspection software (e.g., GOM Inspect), provides a more comprehensive description of surface trueness and overall geometric accuracy [14].

From the analysis of existing research, it becomes evident that while many studies address partial aspects of reverse engineering accuracy, there is no unified, repeatable, and standardised methodology for evaluating the accuracy of reverse engineering systems as a whole [12].

This gap motivates the need for a methodological framework that:

- defines a suitable test artifact;
- specifies measurement procedures for both physical and digital models;
- standardises the evaluation process;
- allows repeatable comparison of different reverse engineering setups.

The following sections therefore propose a standardised methodology for accurately assessing reverse engineering systems and experimentally verifying their applicability using a complete CAD–print–scan–reconstruct cycle.

2. MATERIALS AND METHODS

When reverse engineering is carried out using 3D scanning, the overall process includes successive stages of physical object digitisation, processing of the acquired data in the form of a polygonal mesh, possible mesh modifications, conversion into a CAD model, and subsequent reproduction or further utilisation of the digital geometry. Each of these stages introduces its own type of geometric error into the process, and the resulting inaccuracy is the mathematical sum of the errors associated with individual steps, while also being strongly influenced by their mutual interactions and parameter settings within the entire workflow.

For this reason, the objective of this work is not only to determine the final error at the end of the process, which

expresses the degree of inaccuracy of the overall reproduction, but also to perform a detailed analysis of the individual stages of the reverse engineering cycle, identify dominant sources of error, and investigate possibilities for their minimisation through appropriate scanner software settings and data post-processing strategies.

In order to ensure accuracy is evaluated objectively, this work adopts an experimental approach in which a reference CAD model created by the author is available at the beginning of the process.

The use of a CAD model as the starting point of the methodology is deliberate and necessary to objectively evaluate the accuracy of the individual process stages. By creating a custom CAD model, it is possible to clearly define the reference geometry and subsequently analyse the accuracy of the manufacturing device, which in the context of reverse engineering represents a common production system, such as a 3D printer.

The first step of the methodology is to design a test specimen in the CAD environment. This specimen serves as a reference object, and its geometry facilitates accuracy evaluation in different directions and across various types of geometric features. The CAD model represents an ideal, theoretical geometry free of manufacturing errors and serves as the basis for subsequent comparisons.

The CAD model is then physically created using 3D printing. At this stage, the first source of geometric deviation is introduced, as additive manufacturing is subject to technological limitations such as axis positioning accuracy, layer resolution, material shrinkage, and part orientation on the build platform, which introduces a level of manufacturing error to the physical model relative to the original CAD model.

In order to quantify this error, the physical model is subjected to direct dimensional measurements of selected characteristic dimensions. In this work, fourteen dimensions are selected, representing different types of geometric features of the specimen. Measuring these dimensions makes it possible to determine the average manufacturing deviations and provides information about the accuracy of the manufacturing device itself.

The next step in the methodology is to digitise the physical model using a 3D scanner. Scanning is performed under various scanner software settings, which may generally include acquisition modes, point density, smoothing, or noise suppression. The result of the scanning process is a polygonal mesh that represents a digital approximation of the physical surface of the specimen.

It should be emphasised that a mesh model is inherently a discrete representation of the surface, composed of numerous small facets and affected by irregularities caused by noise, interpolation, and data processing algorithms. Consequently, further mesh modifications such as smoothing, simplification, or hole filling can significantly influence the resulting geometry of the model.

At this stage, the methodology considers two possible paths for further data processing. Following the 3D scanning process and the generation of the STL model, two possible workflows can be considered. The first workflow

involves the direct use of the STL file for manufacturing a new specimen without any further geometric modifications. This path represents a situation in which the objective of reverse engineering is the rapid reproduction of an object without the need for further modifications. In such a case, however, the resulting specimen carries all the irregularities and errors inherent in the polygonal representation.

The second option is to convert the polygonal mesh into a CAD model, that is, into a solid file. This step not only enables further modification of the specimen but also leads to a 'regularisation' of the geometry, as the CAD model consists of a smaller number of ideal surfaces. The resulting surfaces are smoother and geometrically more regular; however, this regularisation may suppress local geometric details and thus lead to deviations from the original physical model.

A key step in evaluating the accuracy of the reverse engineering system is comparing the digital data with the reference CAD model. This comparison is performed in inspection software using an alignment function, which compares the scanned model with the reference geometry.

Since the scanned object is already a physical model affected by manufacturing error, the resulting deviation after alignment simultaneously includes the 3D printing process error and the scanning process error, which makes it necessary to separate these errors within the methodology. In this work, this separation is achieved via direct physical measurement of the dimensions of the manufactured reference model, which provides information about the manufacturing error. By subsequently comparing the CAD model with the scanned data, the combined error caused by printing and scanning can be determined. After subtracting the manufacturing error, the error of the scanning process itself can be approximated.

The proposed methodology thus makes it possible to systematically evaluate the accuracy of individual steps of the reverse engineering process as well as the accuracy of the entire system as an integrated workflow, and it establishes a framework within which different software settings, data processing strategies, and manufacturing approaches can be compared, while ensuring that the results are mutually comparable and repeatable.

The experimental research consists of multiple steps, which are discussed in detail separately in the following sections. At the outset, it was necessary to approach the experimental problem in a highly comprehensive manner in order to design a component with an optimal shape and dimensions that would satisfactorily fulfil its intended purpose. This means that the component has to consist of various shapes and dimensions that will effectively demonstrate the accuracy of reverse engineering systems.

After the component had been designed and modelled in the CATIA V5 software environment, six identical specimens were manufactured using a Prusa 3D printer based on FDM (Fused Deposition Modelling) technology. All specimens were produced using the same printing pa-

rameters to ensure the repeatability of the manufacturing process. Once printed, all specimens were scanned using a 3D scanner based on structured-light technology. Subsequently, all six scans were saved in STL-format data files and processed and converted through several intermediate steps into PART-format models using the SolidWorks software.

For comparison purposes, both the original scan in STL format and the final reconstructed PART model were again printed in six copies each. Finally, all selected dimensions were measured on the initial printed parts, the scanned models, and all twelve final printed specimens. Each of these steps involves many operations and parameter settings, which are described in detail, as they directly influence the final results.

A significant portion of the experiment was carried out within various software environments, including component modelling, scanning, scan editing, data conversion, and subsequent dimensional measurement. Throughout the entire experiment, the functionalities of all used software tools were examined, and results obtained under different parameter settings were compared with the aim of optimising the overall process to achieve the highest possible accuracy while considering time efficiency and experimental complexity.

During the preliminary phase of the study, the geometry of the test component was iteratively refined in order to achieve a suitable balance between geometric complexity and reliable data acquisition. The final design was selected to ensure stable and repeatable scanning conditions while still containing a sufficient variety of geometrical features for accuracy evaluation.

The ExScan Pro software provides a wide range of scanning parameters, the influence of which on the scanning results was investigated experimentally by scanning the printed component under various parameter combinations. The resulting scans were subsequently compared in the GOM Inspect software. This comparative analysis played a key role in identifying and selecting the optimal scanning parameter settings for the purposes of this experiment. A concise schematic representation of the experimental process is shown in Figure 1.

This concise diagram defines the experiment and describes its steps. These steps are discussed in greater depth in the following tree diagram of the experimental plan, which also expands on them with additional, previously unmentioned steps. Within the 3D-scanning process diagram, the progressive development of the software parameter settings is illustrated in Figure 2, including:

- pre-scanning settings (non-HDR / HDR) – option to perform scanning in HDR quality;
- post-scanning settings (HIGH detail / LOW detail) – preservation of details during the conversion of the point cloud into a mesh;
- mesh processing (non-SMOOTH / SMOOTH) – option to apply surface smoothing to the scan;
- mesh processing (non-SIMPLIFIED / SIMPLIFIED) – option to reduce the number of polygons.

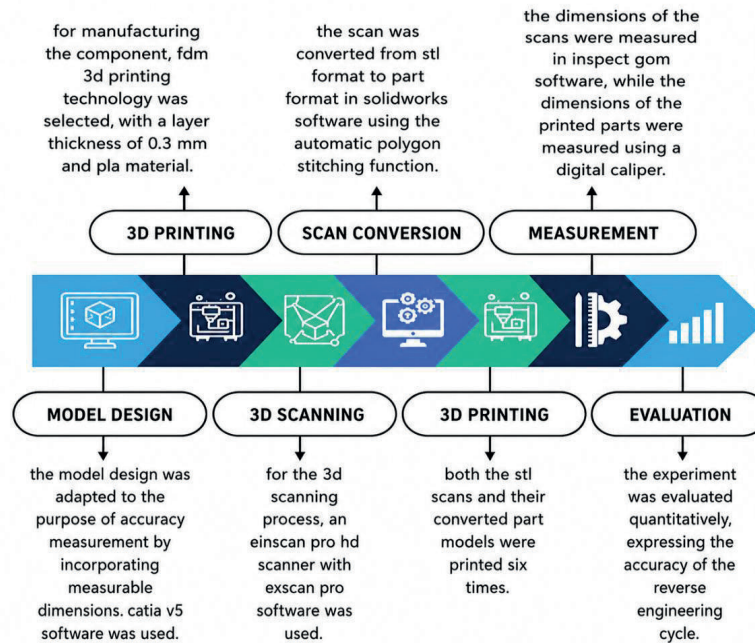


Fig. 1. Concise schematic representation of the experimental process

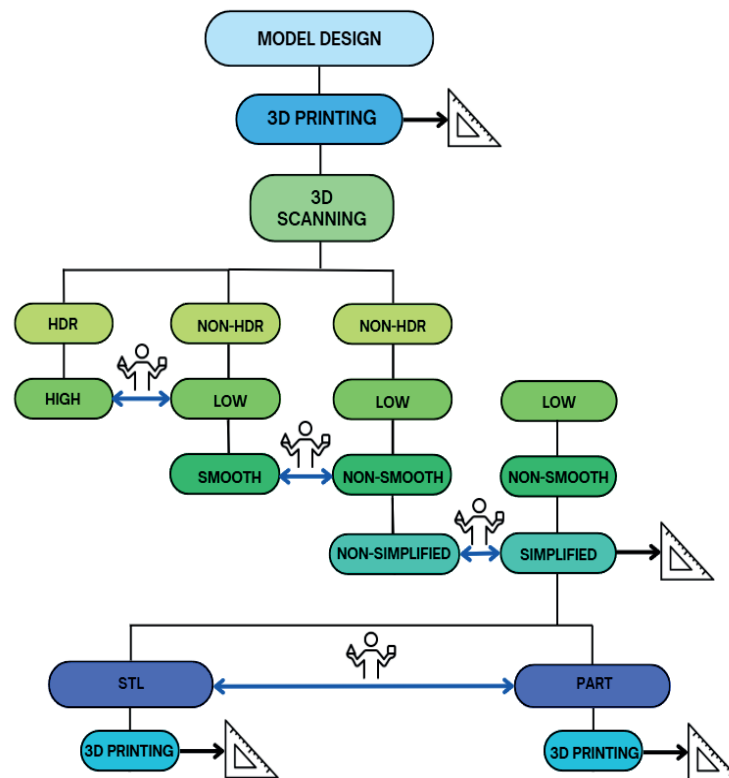


Fig. 2. Tree diagram of the experimental plan

By progressively comparing the scans, as indicated by the bidirectional blue arrow, decisions were made regarding which parameter settings would be used for the experiment. The comparisons were carried out in the Inspect GOM software using the deviation colour map function, which was used to visualise the extent to which individual scanning and post-processing settings influenced the dimensional accuracy and surface trueness of the resulting scan. The final column

was selected as the optimal configuration, namely scanning without HDR, mesh generation with the lowest quality setting (LOW), no surface smoothing (NON-SMOOTH), and, in order to reduce the file size, mesh simplification to 4% of the original polygon count. All scans were then printed in their original STL format as well as in their converted PART format versions. All components (printed parts and scans) whose dimensions were measured are marked with a ruler symbol.

2.1. Design of the test artifact

The first and most critical step of the experiment is the design of a suitable test artifact, developed by the author specifically for the purposes of this study, as the ability to demonstrate and evaluate the accuracy of reverse engineering systems directly depends on the properties of the selected model. The geometry of the artifact must satisfy conflicting requirements: it should not be excessively complex in order to avoid complications during additive manufacturing and, in particular, during 3D scanning, while also being sufficiently diverse to demonstrate the capability of the 3D printer to fabricate smaller and more intricate features, and the ability of the structured-light 3D scanner to capture such features with adequate detail.

The CATIA V5 CAD software was used to design the test artifact. The model was conceptually divided into two parts, which are evaluated separately. The main part consists of a cuboid with dimensions of 80 mm × 40 mm × 10 mm, as shown in blue in Figure 3, while the second part comprises six smaller geometrical features located on the top surface of the main body, as shown in yellow in Figure 3. This division reflects the fact that both the 3D printer and the 3D scanner exhibit significantly different accuracy depending on the size and shape of the manufactured or scanned object.

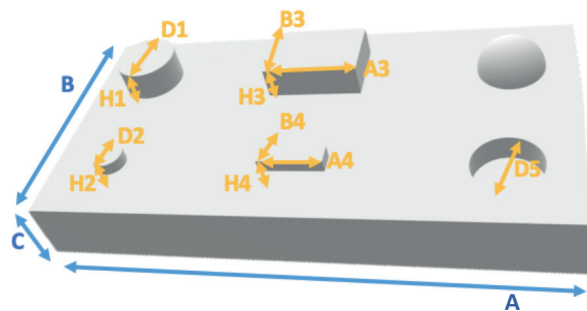


Fig. 3. Designed test artifact

For this reason, the accuracy of the two parts is assessed independently. The first evaluation focuses on the accuracy of the 3D printing and 3D scanning of a simple prismatic geometry with flat surfaces and sharp edges. The second evaluation investigates the accuracy and capability of the 3D printing and 3D scanning processes when applied to small features of different geometrical characteristics, including prismatic, circular, and spherical shapes. This approach makes it possible to directly compare the performance of the reverse engineering system when dealing with simple reference geometry and when capturing and reproducing fine and geometrically diverse features. Figure 3 shows all measured dimensions divided into two categories, as previously mentioned.

Table 1
Dimensions of the designed artifact

Designation	A	B	C	D1	H1	D2	H2	H3	A3	B3	H4	A4	B4	D5
Reference value (mm)	80	40	10	10	5	5	2	5	15	10	2	10	3	10

The numbering of the dimensions corresponds to the individual geometrical features of the test artifact rather than to consecutive dimension identifiers. Therefore, some indices are intentionally omitted.

Table 1 shows all dimensions of the designed artifact.

2.2. Manufacturing of the test artifact

The second step of the experimental methodology is to additively manufacture the designed test artifact using an Original Prusa i3 MK3 3D printer based on the FDM technology. Prior to initiating the 3D printing process, the CAD model created in CATIA must be exported correctly into the STL file format. Incorrect export settings can lead to improper slicing of the model, which subsequently results in inaccurate layer generation and printing errors. After converting the CAD model into an STL file, the file is imported into the PrusaSlicer software, which is freely available and commonly used in combination with 3D printers manufactured by Prusa Research. The printing parameters are defined within the slicing software prior to manufacturing. The test artifact was printed six times from PLA filament using a layer height of 0.3 mm and a nozzle diameter of 0.4 mm. An infill density of 15% was selected, resulting in an approximate printing time of 45 minutes per specimen.

2.3. 3D scanning of the test artifact

The scanning process represents a complex and highly flexible part of the reverse engineering workflow. It can be broadly divided into two interconnected domains: external conditions and software-based processing. External factors, such as lighting conditions, surface properties of the scanned object, scanning angles, and object positioning, can be influenced to a certain extent and were adjusted in this work in order to achieve the most stable and repeatable scanning conditions possible. Prior to scanning, a thin layer of white matte scanning spray was applied to the surface of each specimen in order to reduce reflections and improve the quality and consistency of the data acquisition. However, an equally important part of the scanning process takes place within the scanner software environment, where the operator plays a crucial role and continuously influences the resulting quality and accuracy of the acquired data.

The scanning was performed using an EinScan Pro HD structured-light 3D scanner (SHINING 3D) operated through the ExScan Pro software. In the Fixed Scan mode used in this study, the scanner provides a single-shot accuracy of 0.04 mm, a point resolution of 0.24 mm, a scanning area ranging from 209 mm × 160 mm to 310 mm × 240 mm, and a scanning speed of less than 0.5 s per scan, according to the manufacturer's specifications. The software provides a wide range of functions and adjustable parameters. Depending

on the operator's knowledge and experience, these settings may lead either to highly accurate results or to significant inaccuracies. For this reason, particular attention was paid to the selection and combination of software parameters. Fixed scanning mode was selected for data acquisition, in which the scanner remains stationary while the object is scanned from multiple viewpoints. Although real scanning environments are rarely ideal, the scanning conditions were adapted to minimise disturbances and ensure consistent data acquisition. These external adjustments were made to support the software-based scanning process and are not the primary focus of this methodology. Within the scanning software, several key parameters significantly influenced the resulting data. One of the main parameters was the number of scanning positions, which determines how many individual scans are captured during one scanning cycle. In this work, eight scanning positions were selected, meaning that the object was scanned at evenly distributed angular intervals.

During the scanning process, the individual scans are progressively aligned and merged, resulting in a gradually forming digital representation of the scanned object. After completing a single scanning cycle, the resulting data consisted of an incomplete point cloud, which could not yet be used for further reverse engineering steps. To obtain complete geometric information, additional scanning cycles were performed with the object repositioned, allowing data acquisition from previously inaccessible surfaces. The software subsequently aligned and merged the individual scans based on the detected geometric features, resulting in a dense point cloud representing the scanned artifact. Camera brightness was another important software parameter, whose intensity directly affected the scanner's ability to capture surface geometry. Proper adjustment of this parameter required operator experience, as it had to be adapted to the prevailing external lighting conditions. Inappropriate brightness settings led to missing data or distorted geometry, particularly in areas with limited visibility or an unfavourable surface orientation.

The resulting point cloud represents the primary digital output of the scanning process and serves as the main input for subsequent reverse engineering steps. The density of the point cloud depends on the selected scanning resolution, with a higher resolution resulting in hundreds of thousands to millions of points. While a higher point density provides more detailed geometric information, it also increases the data complexity and computational demands in later processing stages.

Post-processing represents the next critical step and is performed within the same scanning software environment. During this phase, the point cloud is inspected and manually corrected where necessary. In particular, sparse or irregular point distributions occurring at sharp edges or transitions between surfaces were identified and removed, as they were considered scanning artifacts with unknown origin Figure 4. If left uncorrected, such irregularities could lead to significant geometric errors in the reconstructed surface.

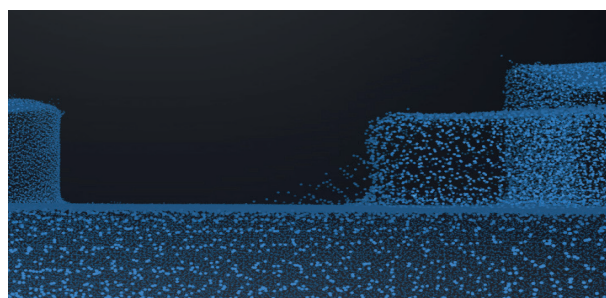


Fig. 4. Sparse or irregular point distributions

After the point cloud was adjusted to a satisfactory state, a polygonal mesh was generated by connecting the points into a large number of triangular facets, resulting in a watertight mesh. The duration of this process and the size of the resulting file depend on the selected resolution level. The initial mesh typically includes all captured data points, except for very sparse regions that are automatically discarded by the software. However, such meshes are often unnecessarily large, making them difficult to handle in subsequent software tools.

Influence and selection of ExScan Pro software settings on dimensional accuracy

The influence of selected scanning and post-processing parameters was systematically evaluated to achieve an optimal balance between dimensional accuracy and data efficiency. The analysis focused on three key aspects: HDR scanning mode, mesh resolution, and post-processing functions, including smoothing and mesh simplification.

The evaluation was performed using comparative deviation analysis in the GOM Inspect inspection software, as shown in Figure 5, where different parameter combinations were applied to the same scanned object and the resulting geometries were compared. The methodology was based on isolating the effect of individual parameters while keeping all other conditions unchanged.

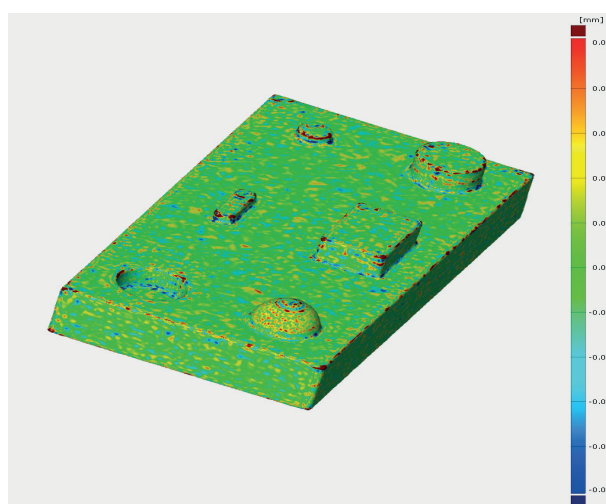


Fig. 5. Deviation map of comparison of non-simplified and simplified scanned models

First, the influence of the HDR mode and mesh resolution was assessed by comparing extreme configurations (HDR

with high resolution versus non-HDR with low resolution). The results showed only minor local deviations, primarily in geometrically complex areas, while no significant differences were observed on planar surfaces. Since the increase in data size and computational complexity was not accompanied by a measurable improvement in dimensional accuracy, the HDR mode combined with high resolution was not considered efficient for further use.

Subsequently, the effect of surface smoothing was evaluated. Although smoothing improved the visual continuity of the mesh, deviation analysis confirmed that it had a negligible influence on the dimensional accuracy. Therefore, this operation was not included in the final workflow, as it increased the processing time without providing measurable benefits.

Finally, the influence of mesh simplification was analysed. The results demonstrated that reducing the number of polygons significantly improved data manageability while preserving dimensional accuracy. No systematic geometric distortion was observed after simplification.

Based on these findings, the optimal configuration for subsequent processing was defined as non-HDR scanning with a low mesh resolution and applied mesh simplification, while smoothing was omitted. This configuration ensures sufficient dimensional accuracy, together with efficient data handling and processing.

2.4. Conversion of the scanned mesh into a CAD model

The direct output of the 3D scanning process is a polygonal mesh in STL format, which represents the object's surface as a collection of interconnected triangular facets. While this representation is suitable for visualisation and certain manufacturing applications, it does not allow the direct parametric modification or advanced geometric analysis typically required in reverse engineering workflows. For this reason, it is often necessary to convert the mesh into a CAD model.

In this work, the conversion was performed using the 'Scan to 3D' functionality available in SolidWorks. This tool transforms polygonal mesh data into a surface-based CAD representation by automatically fitting analytical and freeform surfaces to the scanned geometry. Mesh simplification was applied prior to conversion to reduce the data complexity and ensure stable processing within the CAD environment.

A deviation analysis was performed to evaluate the influence of this conversion step on the dimensional accuracy by comparing the original mesh with the resulting CAD model. The deviation map revealed that most of the surface remained within a narrow tolerance range, indicating a high level of geometric fidelity between the mesh and the reconstructed CAD model. Localised deviations were observed primarily in regions with high curvature and sharp geometric transitions, where surface approximation is inherently more challenging.

Figure 6 presents a zoomed-in deviation map comparing the mesh with the converted CAD model, confirming that the mesh-to-CAD conversion preserves the overall geometry with only minor localised deviations reaching a value 0.32 mm, which may further complicate remanufacturing the model.

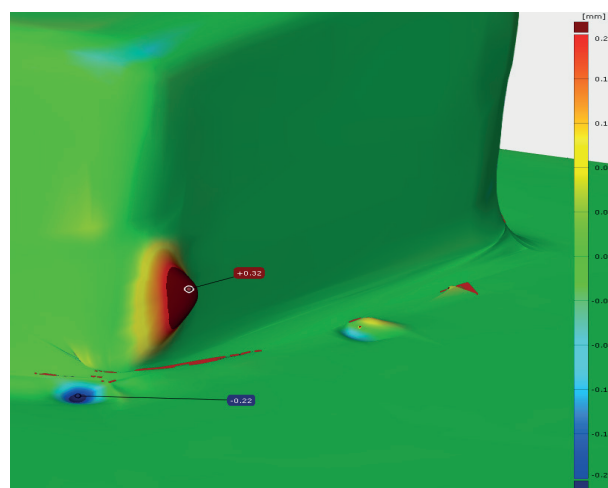


Fig. 6. Deviation map comparing the mesh and the converted CAD model

2.5. Manufacturing and evaluation of the reconstructed model

Following the mesh-to-CAD conversion, both the original scanned mesh and the reconstructed CAD model were physically reproduced using the same additive manufacturing process. This step made it possible to directly compare their manufacturability, surface quality, and dimensional fidelity.

Both digital representations were prepared under identical conditions and subsequently fabricated in order to evaluate how different data representations influence the final physical output. This comparison is essential, as reverse engineering workflows can use either direct mesh-based reproduction or CAD-based reconstruction depending on the application.

The comparison confirmed that both printed models exhibit surface irregularities. In the mesh-based model, these originate from inherent imperfections in the scanned data, which are directly preserved in the polygonal representation. In contrast, the reconstructed CAD model shows additional surface deviations caused by geometric approximation introduced during the conversion process.

It was also observed that even after reconstruction, the resulting geometry does not fully eliminate surface imperfections, and new local deviations may appear as a consequence of surface fitting. This indicates that the conversion step itself represents a potential source of geometric inaccuracy.

2.6. Dimensional measurement and accuracy evaluation

Dimensional measurements of the initial printed specimens were performed using a digital calliper. Each characteristic dimension was measured six times, and the arithmetic mean of the six measurements was used for the subsequent evaluation. The obtained values were then compared with the reference CAD model created in CATIA V5. This comparison was used to quantify the manufacturing errors introduced by the additive manufacturing process. Subsequently, each scanned model was manually measured in the GOM Inspect inspection software, where the same fourteen characteristic dimensions were evaluated directly on the digital geometry.

The measured digital dimensions were then compared with the corresponding dimensions obtained from the physical printed specimens. This comparison enabled the quantification of the dimensional accuracy of the 3D scanning process independently of the manufacturing error.

Digital comparison of the mesh and the reconstructed CAD model using deviation maps in the GOM Inspect software provides information about the errors introduced during the mesh-to-CAD conversion process in SolidWorks. Furthermore, the inaccuracy introduced in the final stage of the process can be quantified by physically measuring the printed reconstructed parts and printed scan-based models using a digital calliper and comparing these measurements with their digital counterparts.

Dimensional measurement of the scanned model

Dimensional measurements were performed directly on the scanned mesh using the GOM Inspect inspection software to evaluate the dimensional accuracy of the 3D scanning process. This software provides advanced tools for measuring geometric features and comparing digital models with reference geometry.

The scanned mesh was first imported into the software environment, where reference geometric elements were created at the locations corresponding to the selected characteristic dimensions. Planar features were generated using the Gaussian best-fit method, which calculates the optimal plane based on the spatial distribution of mesh points while minimising the influence of local irregularities. This approach reliably extracts dimensional information from non-ideal scanned surfaces.

For cylindrical features, reference cylinders were generated using the same best-fit principle. The diameters of these features were then calculated directly from the fitted geometric elements. Figure 7 shows the creation of reference geometric features used for the dimensional measurement of the scanned model. For cylindrical features, reference cylinders were generated using the same best-fit principle, and their diameters were calculated directly from the fitted geometric elements.

The original mesh visualisation was temporarily hidden to ensure accurate and unobstructed measurements, allowing direct inspection of the fitted reference geometry, which is represented in Figure 8.

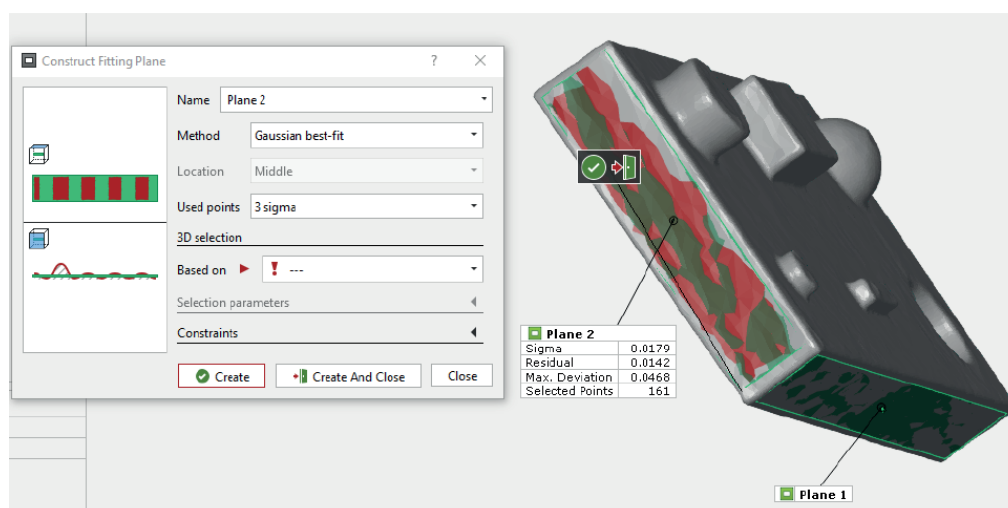


Fig. 7. Plane creation with 'Gaussian best fit' function

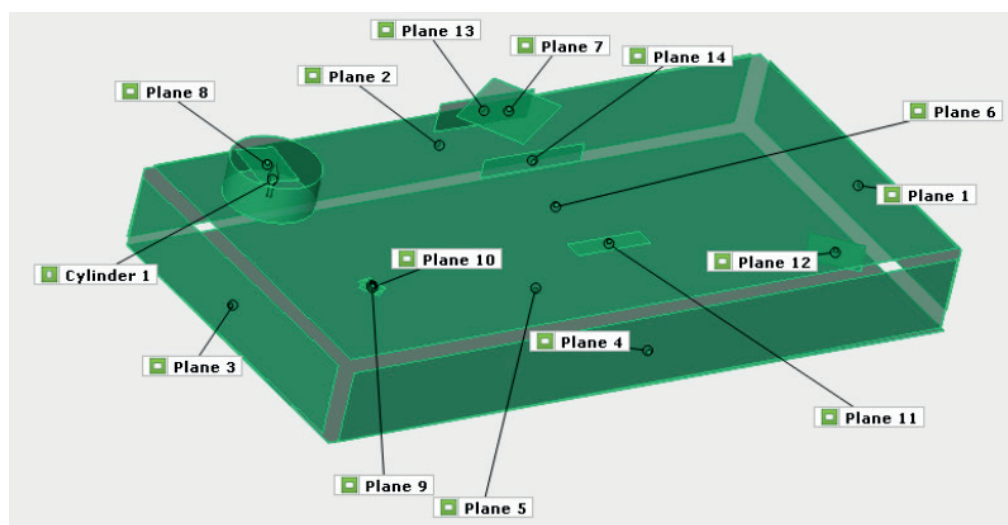


Fig. 8. All created surfaces on the hidden mesh model

Dimensional measurements were performed using the Construct → Distance → Point and Plane function in the GOM Inspect software. This function measures the perpendicular distance between two opposing surfaces. The first point was defined on one surface of the model, and the corresponding point on the opposite surface was automatically projected along a direction normal to the reference plane, ensuring that the measured distance represented the true perpendicular dimension.

For each evaluated dimension, six independent distance measurements were performed at approximately uniformly distributed locations across the measured surfaces, and the resulting values were recorded (see Fig. 9).

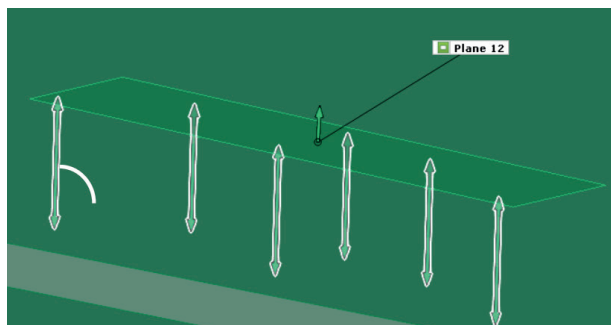


Fig. 9. Measurement method of the scanned model dimensions

3. RESULTS

The evaluation of the experimental results was performed by analysing individual stages of the reverse engineering cycle separately in order to quantify their respective contributions to the overall dimensional inaccuracy. The analysis followed the complete workflow, starting with additive manufacturing of the reference model, continuing through 3D scanning and mesh processing, and concluding with the fabrication of the reconstructed model.

The dimensional accuracy at each stage was determined by comparing the measured characteristic dimensions between consecutive steps of the process. Specifically, the accuracy of the additive manufacturing was evaluated by comparing the dimensions of the printed reference specimen with the original CAD model. The accuracy of the 3D scanning process was determined by comparing the dimensions measured on the scanned model with those obtained from the physical printed specimen. Finally, the accuracy of the reconstructed model reproduction was evaluated by comparing the dimensions of the final printed models with their corresponding digital representations.

Figure 10 illustrates the evaluation scheme used to quantify the dimensional accuracy at each stage of the reverse engineering workflow. The individual differences between the reference dimensions, printed dimensions, scanned dimensions, and reconstructed model dimensions represent the isolated inaccuracies introduced by the additive manufacturing, 3D scanning, and subsequent reconstruction and reproduction steps.

In addition to dimensional deviations, surface quality was evaluated using statistical analysis of the measured values. The standard deviation of repeated measurements

was used as an indicator of surface consistency and local geometric stability. All selected characteristic dimensions were included in the analysis to ensure a comprehensive evaluation of the entire specimen geometry.

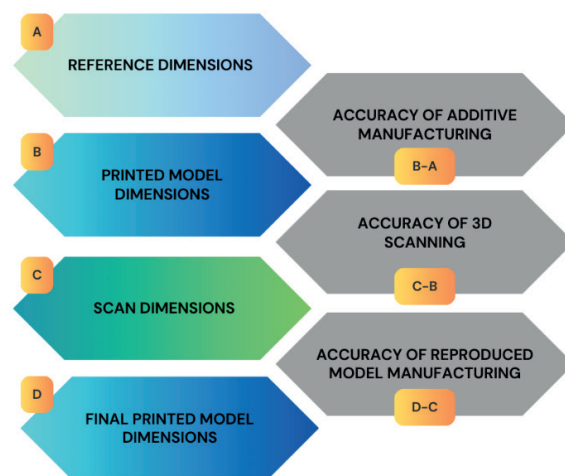


Fig. 10. Evaluation scheme for quantifying dimensional accuracy

3.1. Dimensional accuracy of the reverse engineering cycle

In this section, all measurements previously performed individually for each stage of the reverse engineering workflow are summarised in the following tables and graphical representation. Since the focus of this study is on the proposed evaluation methodology rather than on the specific accuracy values of individual reverse engineering systems, the presented results serve primarily to demonstrate the applicability and structure of the methodology rather than to provide a detailed performance assessment of particular devices or technologies.

Table 2 summarises the average deviations associated with the individual stages of the reverse engineering workflow, representing the isolated accuracy of additive manufacturing, 3D scanning, mesh-based reproduction, and CAD-based reconstruction. These values were obtained by averaging the deviations of the measured characteristic dimensions presented in the previous sections.

Table 2

Isolated dimensional deviations of all the reverse engineering steps

Process type	a	b
3D printing	0.04	-0.05
3D scanning	-0.16	-0.11
3D printing of scan	0.00	0.05
3D printing of part	-0.06	-0.09

Table 3 further presents the dimensional deviations of individual models relative to the original reference CAD geometry. Unlike the isolated deviations of individual processes, these values represent cumulative deviations, as each stage builds on the previous one and propagates its associated inaccuracies.

Table 3

Dimensional deviations of individual models relative to the original reference CAD geometry

Process type	a	b
3D printing	0.04	-0.05
3D scanning	0.12	-0.11
3D printing of scan	-0.12	-0.06
3D printing of part	-0.18	-0.09

Figure 11 graphically illustrates the progression of dimensional deviations relative to the reference geometry throughout the reverse engineering cycle. It can be observed that dimensional deviations increase after the scanning stage and remain present in the final reproduced models, confirming the cumulative nature of dimensional errors introduced during the workflow.

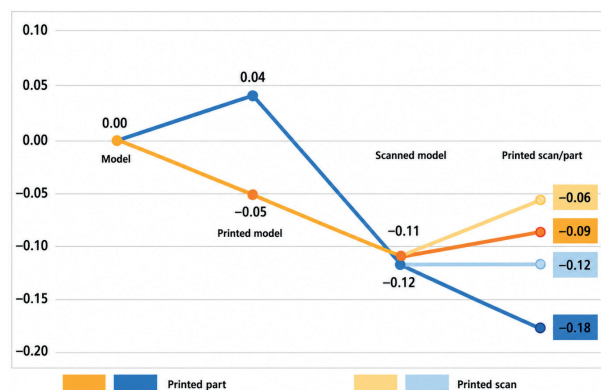


Fig. 11. Progression of dimensional deviations relative to the reference geometry throughout the reverse engineering cycle

The final dimensional deviations depend on whether the reproduced model was manufactured directly from the scanned mesh or from the reconstructed CAD model. With reproduction using the reconstructed CAD model (mesh-to-CAD conversion), the deviations reached approximately -0.18 mm for dimension group 'a' and -0.09 mm for dimension group 'b'. In contrast, when the scanned mesh was reproduced directly without conversion to a CAD model, the deviations were lower, reaching approximately -0.12 mm for dimension group 'a' and -0.06 mm for dimension group 'b'. These results indicate that the mesh-to-CAD conversion step introduces additional geometric deviations, which contribute to the overall dimensional inaccuracy of the reverse engineering cycle.

3.2. Surface quality evaluation of printed models

In addition to the dimensional accuracy evaluated in the previous sections, it is necessary to consider the variability of individual measured values, which reflects the surface irregularities of the printed models. This characteristic can be quantified using the standard deviation of all measured dimensional values for the printed reference models, the printed scan-based models, and the printed reconstructed CAD models. Six repeated measurements were performed

for each characteristic dimension, and the arithmetic mean was used for the dimensional evaluation. The standard deviation was calculated from these repeated measurements to quantify the variability of the measured values and to assess the local surface consistency of the printed specimens.

The standard deviations of the printed reference models, printed scan-based models, and printed reconstructed CAD models are graphically illustrated in the following Figure 12.

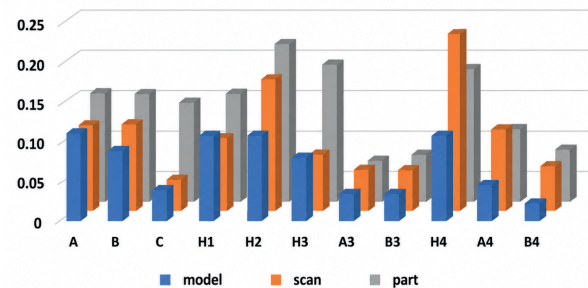


Fig. 12. Standard deviation comparison for three types of manufactured parts

The graphical comparison of the standard deviation values for all three printed models clearly demonstrates a gradual increase in the average standard deviation from the reference model print to the reconstructed CAD model print. This trend is consistent with the irregular surface characteristics of the STL-based scanned model, which are directly transferred to the scan-based print. These surface irregularities also influence the reconstructed CAD print; however, the higher standard deviation observed for the reconstructed model is attributed to additional surface distortions introduced during the conversion process in SolidWorks.

4. DISCUSSION

The authors of this study presented a systematic experimental evaluation of dimensional accuracy within a complete reverse engineering workflow, encompassing CAD design, additive manufacturing, 3D scanning, mesh processing, CAD reconstruction, and final reproduction. The proposed methodology enabled the separate assessment of individual process stages as well as analysis of their cumulative influence on the final dimensional fidelity of the reproduced object.

The results demonstrate that dimensional deviations progressively accumulate throughout the reverse engineering cycle. Each stage of the workflow (manufacturing, scanning, data processing, and reconstruction) introduces its own contribution to the overall inaccuracy. Direct reproduction of scanned mesh data results in lower cumulative deviation compared to reproduction via mesh-to-CAD conversion, indicating that automatic surface reconstruction introduces additional geometric approximation effects.

Surface quality evaluation further confirmed that the geometric consistency decreases with each additional

processing step. The increased variability observed in the reconstructed models highlights the influence of both scanning irregularities and the surface-fitting procedures applied during CAD conversion.

The presented evaluation framework provides a structured and repeatable approach for assessing reverse engineering systems as integrated processes rather than isolated devices. The findings emphasise the importance of considering cumulative error propagation when reverse engineering workflows are applied in dimensionally sensitive applications.

Future research may focus on optimising surface reconstruction strategies and further refining data processing parameters in order to minimise geometric deviations within the complete reverse engineering cycle.

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Modification of Eutectic Silicium in Al-Si-Cu Cast Alloy

Peter Blaško^a , Jozef Petřík^{a*} , Lenka Girmanová^a , Marek Šolc^a , Patrik Fedorko^a 

^a Technical University of Kosice, Faculty of Materials Metallurgy and Recycling, 1/9 Letná St., 042 00 Košice, Slovakia

*email: jozef.petrik@tuke.sk

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Abstract

The bearing alloy AlSi10 with Cu was developed as a possible replacement for aluminium alloys with Sn and Pb. This alloy is intended for low-speed bearings operating under high pressure, which are mainly used in freight transport. An addition of strontium is often used to modify the morphology of eutectic silicon, which affects the properties of the casting. The report analyses its concentration in the alloy, as well as its changes depending on the time between its application to the melt and the actual casting process. Strontium contents above 0.05% lead to a fine and fibrous (globular in the plane of the cut) Al-Si eutectic. The alloy achieved optimal mechanical properties at a strontium content of 0.12%. Hard and brittle needles of SrAl_2Si_2 appear at a strontium content above 0.35%, with their higher content leading to a decrease in strength, plastic properties, and fluidity. On the other hand, they also increase the hardness and practically do not affect the notch toughness. This ‘pre-modification’ is important when the melt is expected to remain at the casting temperature for a longer period.

Keywords:

Al-Si-Cu alloy, eutectic silicon, modifier, properties of casting

1. INTRODUCTION

Al-Si alloys (silumins) with good foundry properties are widely used for the production of all types of sand and die castings. Al-Si-Cu ternary alloys are aluminium casting alloys that use copper to increase their strength and hardness. These can be optimised by subsequent heat treatment through the exclusion of the Al_2Cu phase. These alloys provide low shrinkage, reducing hot tearing. They are also characterised by excellent fluidity and reduced thermal expansion. A copper content up to 5% improves the mechanical properties and machinability, especially after heat treatment, which increases the hardness [1, 2].

On the other hand, copper also has a negative effect: it impairs the corrosion resistance, weldability, ductility, and casting properties, and at higher contents, as an antagonist of silicon, it decreases the fluidity. A higher copper content can impair the resistance to cracking during solidification. It negatively affects the replenishment of the melt into the mould, which can lead to an increased incidence of microporosity. As an antagonist of aluminium, it increases the specific gravity of castings [3, p. 93, 4, p. 338].

Al-Si-Cu ternary alloys are suitable for hardening due to the ternary eutectic Al- Al_2Cu -Si with a melting point of 525°C. The solution annealing is carried out at 515 ±5°C with a sufficient holding time to ensure complete dissolution, usually 3–12 hours, followed by quenching and ageing at 150–170°C [5].

Silumins containing copper are used primarily for the production of complex-shaped castings where high hardness and strength are required. They are suitable for the production of leak-proof castings such as petrol engine fittings and carburettors, but also cylinder heads, pistons, pump housings, and complex castings requiring strength. Silumins containing 1% copper are suitable for the production of plain bearings as a replacement for Al-Sn-X and Al-Pb-X alloys [6]. They are ideal where a combination of the good casting properties of silicon and the strength of copper is required.

Alloys are classified by their silicon content, which determines their suitability for various casting methods. Alloys with a high silicon content (e.g., Al-10Si-2Cu) are suitable for die casting, where high fluidity is required. Conversely, alloys with a lower silicon content and a higher copper content (e.g., Al-3Si-4Cu) are suitable for gravity casting in sand and in metal moulds. This includes Alloy 319 (Al-6Si-3.5Cu); the ‘workhorse’ of the automotive industry, used for engine blocks and cylinder heads.

For a given composition of the alloy, it is the ability to optimise the grain size, eutectic structure, cell size, casting soundness, and the size and distribution of intermetallics that determines the properties. The morphology of the eutectic silicon has an important influence on the mechanical properties of Al-Si alloys.

Silicon is a faceted phase and makes the Al-Si eutectic an irregular eutectic. The brittleness of Si-crystals, of course, is the main characteristic responsible for the poor properties of silumins since the large eutectic silicon particles lead to

premature crack initiation and fracture in tension conditions. It has been proven that modification of eutectic silicon plays an important role in improving the mechanical properties of Al-Si alloys. Therefore, many efforts have been made in the modification of casting Al-Si alloys to achieve a fine silicon phase with beneficial shapes and distribution [7].

The process of crystallisation of eutectic silicon (β -phase) in silumins leads under normal conditions to the exclusion of plate-like crystals. Its morphology has a significant impact on the values of the mechanical properties. Changing the shape of the eutectic silicon (modification) is based on changing its growth unit from a plate to a rod (fibrous, in the plane of the cut from a needle to a globule). The result is a change in the fracture micromechanism from cleavage of the brittle β phase to ductile fracture of the aluminium α matrix. This change is associated with an increase in the strength and plastic properties of the alloy. It can occur chemically: by adding a modifier, using a higher cooling rate, or overheating the melt. The literature provides a wide range of modifiers, such as Li, Ce, Be, Na, Ca, Rb, Sr, As, Sb, Pb, and Bi.

Due to the limited effectiveness, price, and, above all, due to the toxicity of most of them, only strontium and sodium are practically used [7].

Na and Sr have by far the strongest modifying effect at very low addition levels. Although similar mechanical properties can be achieved by adding Sb, the effect of Sb on the microstructure of the eutectic is significantly different from that of Na or Sr. The addition of Sb refines the Al-Si eutectic, producing a fine flaky structure instead of the fibrous structure produced by Na or Sr [8, 9].

Sb is used mainly in Europe and Japan and is commonly referred to as a permanent modifier [10]. Unlike Sr and Na, once added to the aluminium melt, it remains a permanent component of the aluminium alloy. The modifier is therefore added by the foundry supplier and does not require any replenishment at the foundry [11]. Despite the good properties of sodium as a modifier [3], we used strontium as a modifier in our study due to Na's dosing problems and uneven concentration in the melt (mainly due to rapid burnout).

According to the literature, there are different opinions on the optimal concentration of strontium used as a modifier. Older authors recommended 0.03–1% strontium in the melt [3], while some current authors consider strontium content above 0.02% undesirable, as it can lead to coarsening of the eutectic silicon and deterioration of the mechanical properties of the casting [12].

The modifying effect of strontium is not permanent. The persistence of the melt at the casting temperature causes burnout: a decrease in the modifier content. According to the literature, the modifying effect of strontium lasts 1–2 hours, but up to four hours has also been reported at an initial content of 0.03%. Burnout can be slowed down by adding 0.05–0.2% beryllium, in which case the modifying effect can be longer than 10 hours [3]. In practical experiments and an initial content of 0.015% Sr in a melt at a temperature of 760°C, its content decreased to 0.012% after five hours. After 24 hours, the melt still contained 0.004% Sr.

The cooling rate has a significant impact on the modification process. A high cooling rate results in the formation of numerous crystallisation nuclei, an increase in the number of eutectic silicon particles, a change in their morphology, and a decrease in the interparticle distance. The disadvantage is the inhomogeneity of the structure with the appearance of 'overmodified islands' with fine eutectic silicon.

Unlike sodium, strontium does not fade as quickly in the melt, making it preferred for industrial applications that require long holding times or low-pressure die casting [13].

The modification, although not as significant as with gravity casting, is also important for high-pressure die casting (HPDC), as described by Brzezinski et al. [14] and Malik et al. [15]. Al-Si alloy can also be modified through the use of mechanical techniques, such as centrifugation, sonic and ultrasonic vibration, squeezing, etc. These mechanical means have been identified to cause grain refinement, density increase, shrinkage, degassing, change in size and shape, and distribution of the second phase [16, 17].

2. EXPERIMENTAL MATERIALS AND METHODS

The experiments aimed to study the influence of the strontium content in the melt, its temperature, and the time from the application of strontium on the microstructure and mechanical properties of the castings, and also the fluidity of the melt. In addition, the possible undesirable influence of the increased strontium content (premodification) on the microstructure and mechanical properties of silumin castings was evaluated. The silumin used contained: 10% Si, 1% Cu, 0.5% Fe, and 0.2% Mn. The strontium content in the castings was at five levels: 0%, 0.025%, 0.07%, 0.12%, and 0.35%. Chemical analysis was performed by the atomic absorption method on a Perkin Elmer 306A spectrometer. The undesirable influence of iron, present as an impurity, on the mechanical properties was compensated by manganese. The alloy was melted in a graphite crucible in an electric resistance furnace. Refining was carried out with Degasser and Tiboral six preparations. The casting temperature was 570°C. The oxide layer (smear) was mechanically removed before each casting.

A lump (up to 5 mm) of AlSiSr10 master alloy containing 8.35% strontium was used as a modifier. The melt was cast without a modifier or 30, 60, 120, and 300 minutes after the application of the modifier. The alloy was cast into an open steel mould, and its cooling rate ranged from $2^{\circ}\text{C}\cdot\text{s}^{-1}$ in the centre of the casting to $20^{\circ}\text{C}\cdot\text{s}^{-1}$ at its edge in contact with the mould. Before casting, the mould had an ambient temperature of 18–22°C.

To determine the fluidity Y_z , part of the melt was cast into a vertical steel mould (harp). A detailed description of the mould and the method of evaluating the fluidity is described by Petrík [18] and Futáš et al. [19].

Uniaxial tensile tests were carried out, and the parameters R_m (ultimate tensile strength), $R_{p0.2}$ (yield strength), Z (reduction of the area /contraction), and A_5 (total elongation) were evaluated. A 200 kN Zwick-Extensometer (Zwick GmbH&Co., Ulm, Germany) was used for both tests. For the tensile test, the mean strain rate $\dot{\epsilon}_{LC} = 0.0004 \text{ s}^{-1}$, and it was

carried out by the 200 kN Zwick-Extensometer according to the ISO 6892-1:2016 standard [20].

The HV10 and HV0.05 hardness tests were performed according to STN EN ISO 6507-1:2023 [21], and the HBW 5/250/30 hardness test was performed according to STN EN ISO 6506-1:2014 [22] on a HPO 250 hardness tester (Fritz Heckert, Leipzig, Germany). A Hanemann hardness tester (Carl Zeiss, Jena, Germany) with a diamond pyramidal indenter with a magnification of 480× fitted to a Neophot 32 optical microscope was used as a measurement device for the micro-hardness HV0.05 measurement. Hardness measurements were performed on polished and etched metallographic surfaces, the production of which is described below.

The impact toughness test in bending (KC and KCV2 Charpy test) was performed on normalised samples, i.e., 10 mm × 10 mm × 55 mm with a 'V' notch depth $c = 2$ mm, and on 10 mm × 10 mm × 55 mm samples without a notch (resp. 5 mm × 10 mm × 55 mm) according to STN EN 10045-1:1990 [23]. A PSW 30 tester manufactured by Prüfmaschine Fritz Heckert, Leipzig, Germany was used.

The testers for macro- and micro-hardness were calibrated using certified reference blocks (CRM) with hardness (H_{CRM}) and standard uncertainty (u_{CRM}), which are listed in Table 1. As can be seen from the results of the calibration, which was carried out according to the EN ISO 6506-2:2025 [24] and EN ISO 6507-2:2023 [25] standards, all the devices used for hardness measurement meet the requirements. The table shows the values determined by the calibration: the repeatability r_{rel} , the error of tester E_{rel} , and the relatively expanded uncertainty of calibration U_{rel} (all in percentages).

Table 1.
Results of calibration hardness testers

Hardness	H_{CRM} [HV]	u_{CRM} [HV]	r_{rel} [%]	E_{rel} [%]	U_{rel} [%]
HBW 2.5/250/30	185	1.65	0.92	−0.27	2.37
HV10	194	1.65	1.13	−0.20	2.16
HV0.05	195	4.00	1.86	−0.42	5.11

Macro- and micro-hardness were measured according to the ISO 6892-1:2016 [20] and EN ISO 6507-1:2023 [21] standards; the average hardness value from five indentations is reported. All hardness measurements were performed by a single operator.

A test load of 0.49035 N (HV0.05), with a 15-second dwell time was applied for microhardness HV0.05. The macrohardness was measured by the Vickers test (HV10) and the Brinell test (HBW 2.5/187.5/30). Manual control of the micro-hardness tester required maintaining consistent indenter penetration rates, which inherently influence the measured hardness values; the speeds fluctuated around $0.5 \mu\text{m}\cdot\text{s}^{-1}$.

The samples for metallographic analysis and hardness tests were cut from the notched toughness test samples with a water-cooled diamond cast into epoxy resin (dentacryl) for consolidation. Metallographic surfaces were prepared by grinding on 120 to 3000 ANSI/CAMI grit papers and subsequently polished with spineline (Al_2O_3)

moistened with water. This was followed by polishing with a diamond paste ($0.5 \mu\text{m}$) moistened with petroleum. The metallographic surface was etched with 25% H_2SO_4 at 75°C in order to make iron-containing particles visible, and subsequently with a 0.5% aqueous solution of HF.

The morphology of eutectic silicon was evaluated using a NEOPHOT 30 optical microscope (Carl Zeiss, Jena, Germany) according to the STN 42 0491:1978 standard [26].

Its interparticle spacing (distance) was evaluated using Formula (1):

$$\lambda_\beta = \frac{1}{M} \times \sqrt{\frac{A}{N}} \quad (1)$$

where:

M – magnification,

N – number of particles in a defined area A .

The particle size of the eutectic silicon was evaluated using the ImageJ software.

3. RESULTS

If the alloy cast from a temperature of 570°C did not contain strontium, eutectic silicon was precipitated in the centre of the casting in the form of isolated needles (lamellae) (Fig. 1), with a width of up to $4 \mu\text{m}$. The increased cooling rate at the edge of the casting was manifested by the shortening of the needles with a slight decrease in the interparticle distance (Fig. 2). When applying 0.07% Sr to the melt and its subsequent casting (after removing the smear, unannealed casting UC), oval silicon particles with a diameter of $1\text{--}3 \mu\text{m}$ with a rare occurrence of needles were precipitated in the centre of the casting (Fig. 3). The difference between the centre and the edge of the casting, in terms of microstructure parameters, is not as pronounced as in the unmodified state. A further increase in the strontium content is minimally reflected in the microstructure: the globular form of eutectic silicon dominates with a decrease in the interparticle distance towards the edges. A strontium content of 0.35% leads to premodification with the exclusion of needle-like SrAl_2Si_2 formations (marked by two arrows), whose microhardness reaches up to 1200 HV0.05 (Fig. 4).

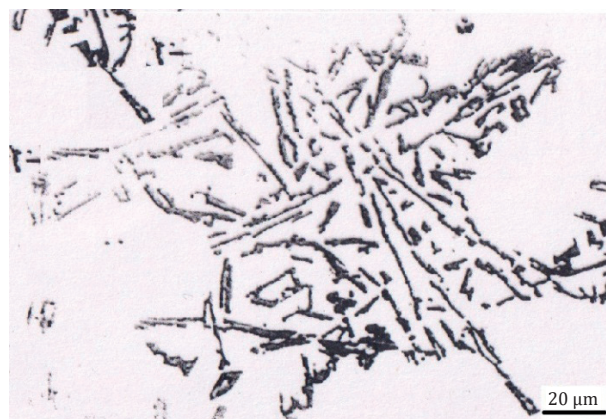


Fig. 1. Alloy without modifier, centre of casting

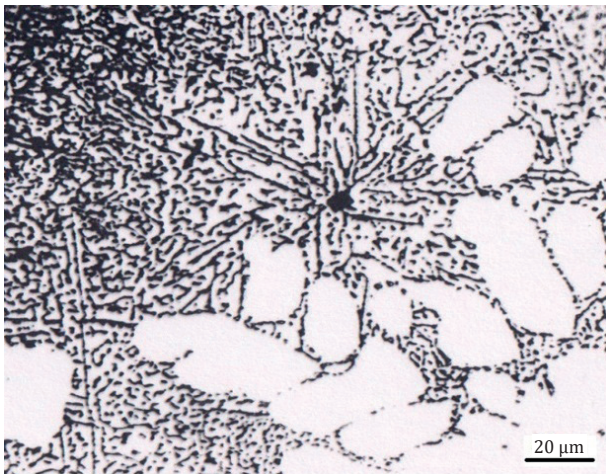


Fig. 2. Alloy without modifier, edge of casting

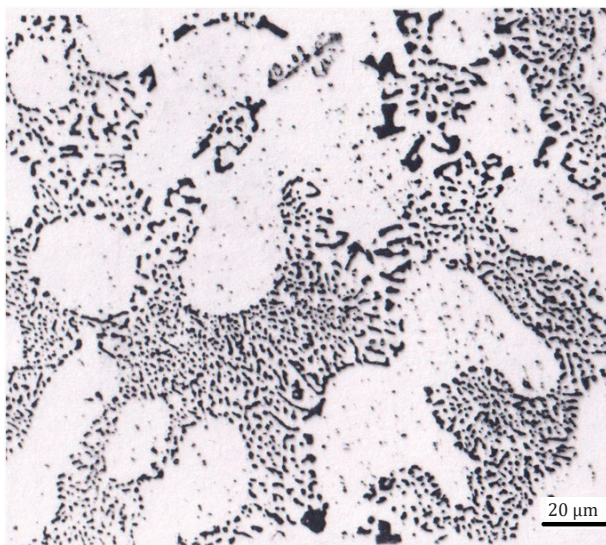
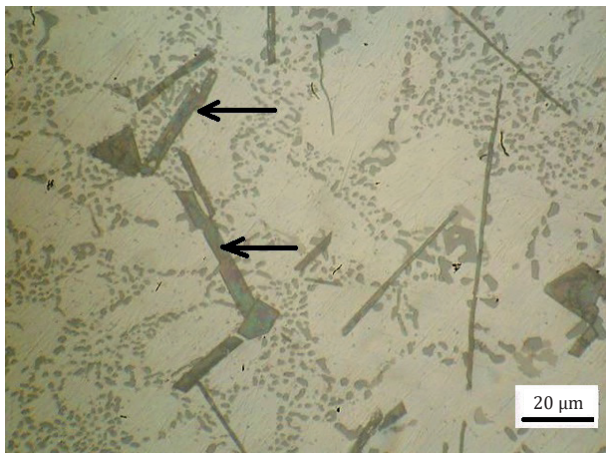


Fig. 3. Structure of an alloy modified with 0.07% Sr

Fig. 4. Precipitation of SrAl_2Si_2 needles after premodification

Figures 5 and 6 show the average values of the strength properties of the samples. Both characteristics increase with an increase in the strontium content up to 0.12%, with a more pronounced effect for R_e , while at higher contents, there is a decrease.

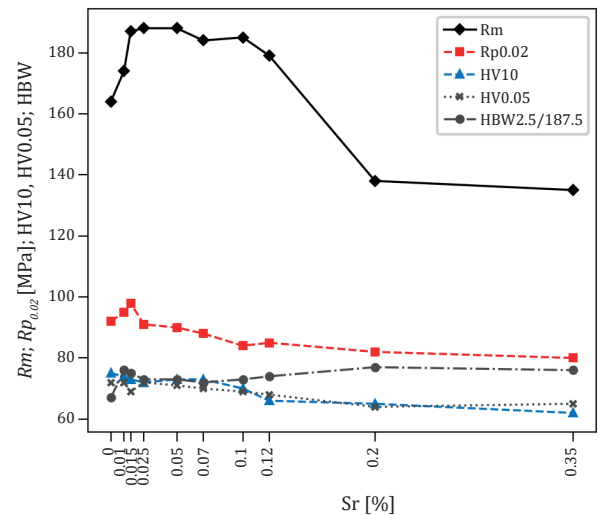


Fig. 5. Effect of strontium content on strength and hardness

Increasing the strontium content leads to a decrease in plasticity, as documented in Figure 5. The decrease at a content of 0.05% is an anomaly, probably due to an influence other than the strontium content. Increasing the strontium content (Fig. 5), leads to an increase in hardness, which is related to the exclusion of hard needle-like formations of SrAl_2Si_2 . The notch toughness reaches maximum values at a content of 0.12% Sr, with a subsequent minimal decrease; a similar trend was also found by the authors [27, 28].

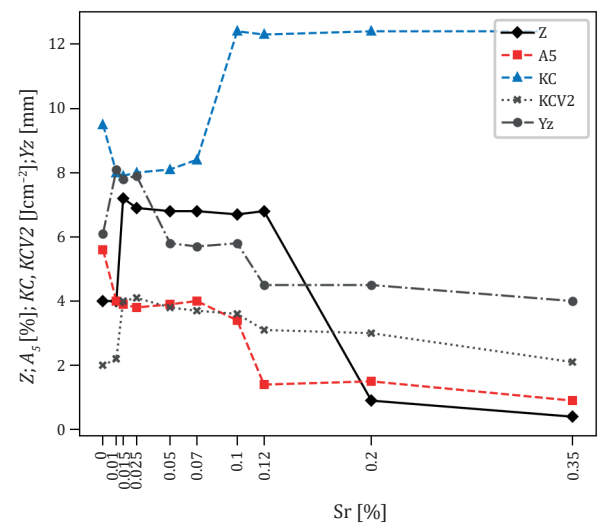


Fig. 6. Effect of strontium on plastic properties, notch toughness/impact value, and fluidity

Increasing the strontium content leads to a decrease in plasticity, as documented in Figure 6. The decrease at a content of 0.05% is an anomaly, probably due to an influence other than the strontium content. Increasing the strontium content (Fig. 5) leads to an increase in hardness, which is related to the exclusion of hard needle-like formations of SrAl_2Si_2 . The notch toughness reaches maximum values at a content of 0.12% Sr, with a subsequent minimal decrease; a similar trend was also found by the authors [27–29].

In the next step, the strontium content and the time between the application of the strontium and the casting were varied. The alloy was cast from a temperature of 570°C into an open mould; the microstructure in Figure 7 is from the centre of the casting.

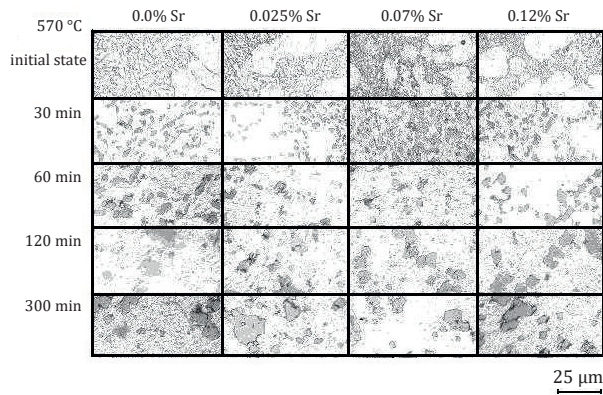


Fig. 7. Effect of strontium content and the time between adding strontium and casting the melt on the microstructure

The simultaneous influence of the initial strontium content and the time between its application and casting on the morphology of eutectic silicon particles is shown in Figure 8 (particle diameter) and Figure 9 (interparticle spacing λ_p).

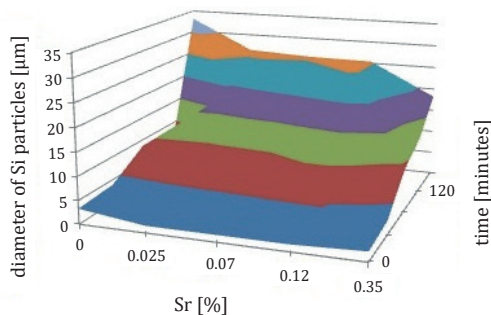


Fig. 8. Effect of strontium content and the time between adding strontium and casting the melt on the diameter of Si particles in the eutectic

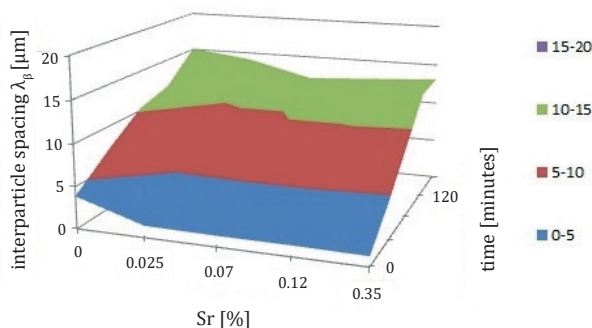


Fig. 9. Effect of strontium content and the time between adding strontium and casting on the melt the interparticle spacing λ_p of Si particles in the eutectic

Figure 10 shows the effect of the simultaneous influence of the initial strontium content and the time between its

application and casting on HV10 hardness. As the time between the application of the strontium and the casting increases, the hardness decreases, which is related to the coarsening of the microstructure. The effect of the initial strontium content (beyond its zero level) is less pronounced.

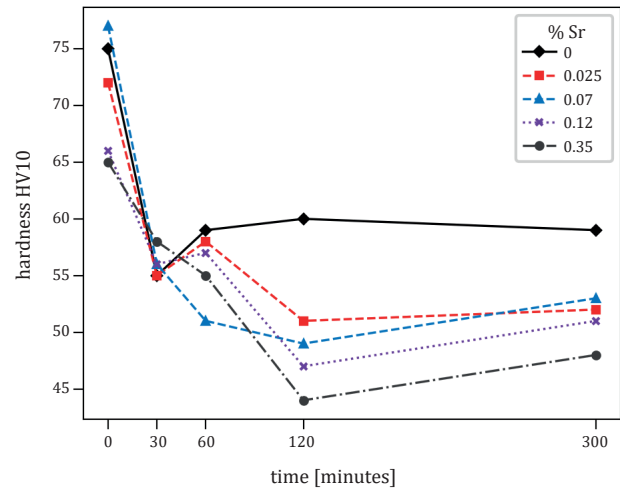


Fig. 10. Effect of strontium content and the time between adding strontium and casting the melt on the HV10 hardness

4. DISCUSSION

The AlSi10 bearing alloy with 1% Cu studied in this work is intended for highly loaded, low-speed plain bearings, where it represents a possible replacement for the traditional Al-Sn-X and Al-Pb-X bearing alloys [6]. In such applications, the combination of strength, plasticity, and castability is decisive, and all of these properties are controlled to a large extent by the morphology of the eutectic silicon. The results of the present experiments confirm that the modification of eutectic silicon by strontium is an effective tool for optimising the microstructure and, consequently, the mechanical properties of this alloy.

In the unmodified state, eutectic silicon precipitated in the form of lamellae (needles in the plane of the metallographic surface), and the alloy exhibited low mechanical properties, particularly the plastic ones (elongation A_5 and reduction of area Z). This observation is in full agreement with the literature: coarse plate-like silicon acts as an internal notch, the fracture proceeds by cleavage of the brittle β phase, and the ductility of the alloy is therefore low. Timpel et al. [30] showed by atom probe tomography and transmission electron microscopy that the addition of strontium changes the growth mechanism of the silicon phase: Sr co-segregates with Al and Si within the eutectic silicon, promotes multiple twinning (impurity-induced twinning), restricts the growth of silicon crystals, and thus transforms the coarse plate-like silicon into a fine fibrous network. In the plane of the section, the fibrous silicon appears as globules, which corresponds to the microstructures observed in our castings at strontium contents of 0.05% and higher.

A strontium content of 0.05% was sufficient for the complete modification of eutectic silicon, and the optimal mechanical properties were achieved at 0.12% Sr, with the most pronounced increase in strength (R_m). The growth of strength and plastic properties associated with the change of the fracture micromechanism from cleavage of the β phase to ductile fracture of the α matrix is consistent with the results of Zamani and Seifeddine [31], who correlated the level of modification (35–500 ppm Sr) with tensile strength and elongation to failure, and with the classical practical analysis of Sigworth [32]. It should be noted that the optimal content found in this work is higher than the values usually recommended for hypoeutectic silumins (approximately 0.01–0.05% Sr [31–33]); this difference can be attributed to the chemical composition of the studied bearing alloy, the relatively low casting temperature (570°C), and the moderate cooling rate in the open steel mould. In accordance with the literature [31, 32], the hardness (HBW, HV) was not significantly influenced by the modifier content, since hardness reflects mainly the volume fractions of the phases present rather than the morphology of the eutectic silicon; a measurable increase in hardness appeared only after the precipitation of hard SrAl_2Si_2 needles.

A further increase of the strontium content up to 0.35% (overmodification) led to the precipitation of brittle needle-like SrAl_2Si_2 phases with a microhardness of up to 1200 HV0.05 and to a decrease in the mechanical properties and fluidity. The same phenomenon is described in the literature: an excess of strontium forms $\text{Al}_2\text{Si}_2\text{Sr}$ intermetallics, which reduce the modification effect, act as internal stress concentrators, and deteriorate the properties of the casting [34]. Emadi et al. [35] further showed that Na and Sr additions decrease the surface tension of the melt and increase the volumetric shrinkage, which supports the observed negative influence of higher strontium contents on castability. The threshold of overmodification found in this work (above 0.35% Sr) is, however, considerably higher than the contents reported as harmful in the literature, which is one of the important practical findings of the present study.

The deliberate overmodification may nevertheless be technologically justified when a longer time interval between the application of strontium and casting is expected. The modification effect of strontium is not permanent: its content in the melt gradually decreases (fades), predominantly by oxidation [35]. Zhang et al. [34] reported that in an A356 melt held at 760°C with an initial addition of 0.06% Sr, the modification effect began to fade after approximately 150 min and disappeared completely after 240 min, when the residual strontium content dropped below approximately 0.008%. Djurdjevic et al. [36] quantified the fading rate and showed that it depends mainly on the melt temperature, the surface-to-volume ratio of the bath, and the furnace size, with sodium fading considerably faster than strontium. A higher initial strontium content therefore acts as a reserve that guarantees complete modification even after longer holding times, which was confirmed by our

experiments with casting 30–300 min after the modifier addition.

Besides strontium, other elements (Na, Sb, Be, Ca, and rare earths) can be used as modifiers. Sodium is the most powerful modifier, but its application is complicated by dosage problems, rapid burnout, and an uneven concentration in the melt [36]. Lu et al. [37] demonstrated that combined Sr + Na additions can be beneficial, but that controlling the sodium content remains difficult in foundry practice. Antimony acts as a permanent refiner, producing a fine lamellar rather than a fibrous structure; it does not fade, but it is less effective, toxic in the form of stibine, and it neutralises the effect of Sr and Na [38]. For these reasons, strontium, added conveniently in the form of an AlSr master alloy with a recovery of up to 90%, remains the preferred modifier for industrial applications, which was also the reason for its selection in this work.

Finally, a modification effect can also be achieved by increasing the cooling rate (quench modification). A cooling rate exceeding approximately $50^\circ\text{C}\cdot\text{s}^{-1}$, typical of thin-walled high-pressure die castings, can refine the eutectic silicon even without a chemical modifier [36]; the influence of the cooling rate on the nucleation and growth of the eutectic was analysed in detail by Dahle et al. [33] and McDonald et al. [39]. This was also confirmed in our experiments by the finer microstructure at the edge of the casting (cooling rate up to $20^\circ\text{C}\cdot\text{s}^{-1}$) compared to its centre ($2^\circ\text{C}\cdot\text{s}^{-1}$). However, the cooling rate is limited by the shape of the casting, the casting method, and the moulding material, and an excessive cooling rate may lead to distortion or even to a loss of integrity of the casting. Chemical modification with strontium therefore remains the most robust and universally applicable method for the studied bearing alloy.

5. CONCLUSIONS

When analyzing the influence of strontium content in an Al-Si alloy on its microstructure and properties, the authors of the paper reached the following conclusions:

1. A strontium content of 0.05% is sufficient for complete modification of eutectic silicon.
2. Optimal mechanical properties were achieved by the alloy at a 0.12% strontium content.
3. A further increase of the strontium content to 0.35% worsens the mechanical properties of the alloy without significantly affecting the microstructure or fluidity.
4. Coarseness of the structure and the beginning of a decrease in mechanical properties were observed at strontium contents above 0.35%, which is higher than reported in the literature.

ACKNOWLEDGEMENTS

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Effect of Cerium Addition on the Evolution of Non-Metallic Inclusions in Steel – A Literature Review

Dorota Kalisz* 

AGH University of Krakow, Faculty of Foundry Engineering, 30 Mickiewicza Av., 30-059 Krakow, Poland

*e-mail: dak@agh.edu.pl

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Abstract

Rare earth elements (REE), particularly cerium (Ce), are widely used as alloying additions in modern steelmaking due to their strong chemical activity and high affinity for oxygen and sulphur. Their ability to react with impurities allows the effective modification of non-metallic inclusions, enhancing steel microstructure and overall performance. This review focuses on the ways cerium interacts in steel, highlighting transformations of inclusions such as Al_2O_3 and MnS , as well as complex systems like Ce-Al-O , Mn-O-S , and Ce-Ca-Al-O . Particular attention is given to the thermodynamic aspects of inclusion evolution and their dependence on processing conditions. The influence of cerium on sulphide stress cracking (SSC) resistance, solidification microstructure, and mechanical properties is also examined. According to the literature, the optimal cerium content is generally 0.01–0.02 wt.%, while higher amounts can negatively affect material performance.

Keywords:

steel, thermodynamic, modification, non-metallic inclusions, sulphides, oxysulphides, cerium

1. INTRODUCTION

Non-metallic inclusions are a major factor limiting steel performance, especially in applications that demand high fatigue strength and corrosion resistance under aggressive conditions. Complete removal of these inclusions from molten steel is not feasible due to technological limitations. Consequently, modern steelmaking focuses on controlling and modifying inclusions to mitigate their negative effects. One of the most effective strategies is the use of rare earth elements, especially cerium. Cerium promotes the transformation of non-metallic inclusions and contributes to improved steel properties. It has a melting point of 799°C and a density of $6,770 \text{ kg/m}^3$ and shows relatively high solubility in ferrite (0.1–0.5%) but low solubility in austenite ($\sim 0.04\%$) [1–3]. In steel, cerium primarily acts as a deoxidiser and desulphuriser, binding low-melting-point impurities such as arsenic, lead, tin, or antimony to form stable compounds including oxides, oxysulphides, sulphides, and nitrides. It also modifies existing inclusions by influencing their chemical composition, morphology, and physicochemical properties, which improves their deformability and enhances the corrosion resistance of the steel. Another important aspect of rare earth elements is their effect on solidification. Small additions of cerium can refine the solid phase and strengthen the steel through phases with limited solubility in the α - γ system. REE additions can also enhance tensile strength by affecting crystallisation and promoting the formation of the

γ -Fe eutectic, which alters the solidification temperature [4]. The effectiveness of cerium in modifying inclusions depends on the sulphur-to-oxygen ratio in the steel and the amount of additive [5, 6]. Research on the fundamental mechanisms of inclusion evolution in REE-modified steels is still limited. The high reactivity of cerium complicates precise metallurgical control, potentially causing issues such as clogging of casting channels during pouring. Cerium also tends to segregate in interdendritic regions during solidification, forming clusters of cerium-rich inclusions. Therefore, its impact is strongly influenced by technological factors such as the deoxidation method, alloy composition, and solidification conditions.

2. THERMODYNAMICS AND KINETICS OF CERIUM REACTIONS IN STEEL

Cerium is a highly reactive element with a strong affinity for oxygen and sulphur, which leads to the formation of stable phases such as Ce_2O_3 , $\text{Ce}_2\text{O}_2\text{S}$, CeS , Ce_3S_4 , and Ce_2S_3 [4–13]. The three-dimensional equilibrium activity diagram of oxygen, sulphur, and cerium (Fig. 1) illustrates how the reactions involving these elements proceed as a function of their concentration ratios. At low concentrations of cerium, oxygen, and sulphur (Ce, O, and S), their activities are strongly influenced by variations in their concentrations. The equilibrium surfaces for cerium oxide (Ce_2O_3), cerium oxysulphide ($\text{Ce}_2\text{O}_2\text{S}$), and the cerium sulphides CeS , Ce_3S_4 , and Ce_2S_3 are shown in Figure 1. The

DEF curve defines the composition of the metal alloy in which cerium oxysulphide ($\text{Ce}_2\text{O}_2\text{S}$) is in equilibrium with the three sulphides. At points E and F, the oxysulphide is in equilibrium with two adjacent sulphides. When the oxygen content in the alloy lies within the equilibrium field of cerium oxysulphide ($\text{Ce}_2\text{O}_2\text{S}$), the addition of cerium results in the immediate precipitation of the oxysulphide. The composition of an alloy containing enough cerium shifts towards the DEF line, along which sulphide precipitation begins simultaneously.

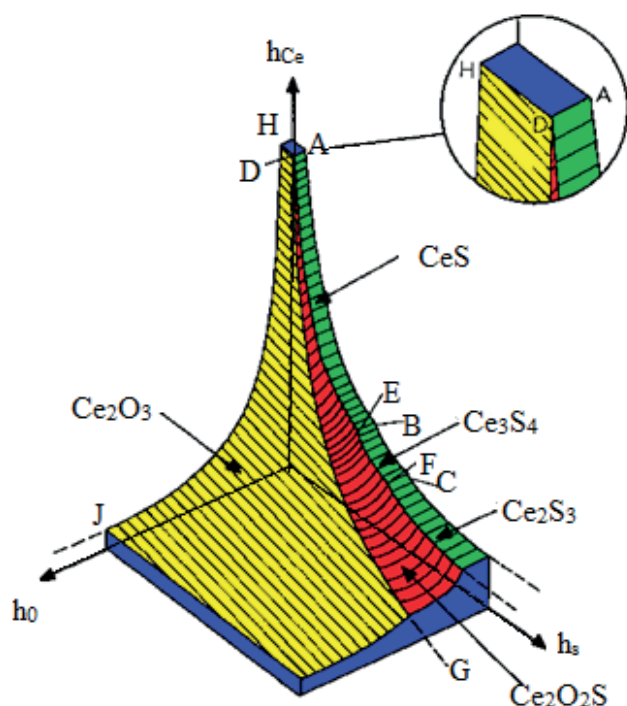


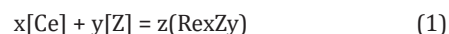
Fig. 1. Ce–O–S equilibrium system [5, 9, 11]

Table 1

Standard Gibbs free energy ΔG° for the reactions of cerium [9, 14–26]

Reaction	Formula no.	$\Delta G^\circ = A+B \text{ [J}\cdot\text{mol}^{-1}\text{]}$
$[\text{Ce}] + [\text{O}] + 1/2[\text{S}] = 1/2\text{Ce}_2\text{O}_2\text{S}(\text{s})$	(4)	$-675,700 + 165.5T$
		$-676,795 + 166T$
		$-675,700 + 164.45T$
$2[\text{Ce}] + 2[\text{O}] + [\text{S}] = \text{Ce}_2\text{O}_2\text{S}(\text{s})$	(5)	$-1,352,700 + 331T$
$[\text{Ce}] + [\text{S}] = \text{CeS}(\text{s})$	(6)	$-394,428 + 121T$
		$-422,100 + 120.38T$
		$-422,780 + 121T$
		$-422,783 + 120.58T$
		$-3,944,428 + 121T$
$[\text{Ce}] + 3/2[\text{O}] = 1/2\text{Ce}_2\text{O}_3(\text{s})$	(7)	$-714,380 + 185.67T$
		$-715,560 + 180T$
$3[\text{Ce}] + 4[\text{S}] = \text{Ce}_3\text{S}_4(\text{s})$	(8)	$-1,494,441 + 439T$
$[\text{Ce}] + 4/3[\text{S}] = 1/3\text{Ce}_3\text{S}_4(\text{s})$	(9)	$-497,670 + 146.37T$
		$-498,480 + 146T$
$[\text{Ce}] + 3/2[\text{S}] = 1/2\text{Ce}_2\text{S}_3(\text{s})$	(10)	$-536,420 + 112.16T$
		$-536,420 + 163.86T$
		$-537,290 + 164T$

Simrnov et al. confirm that Ce can be added to molten steel over a broad concentration range at low Ce concentrations in liquid steel (up to 0.1%), because cerium primarily reacts with oxygen and sulphur [1]. Higher additions promote the formation of oxide and sulphide phases and can also neutralise other detrimental elements such as phosphorus and copper, as well as interact with nickel and iron. The final composition of non-metallic inclusions is strongly influenced by the oxygen and sulphur content in the steel. The formation of inclusions occurs according to the following reaction Scheme (1). The equilibrium constants for the reaction are expressed by Equations (2) and (3) [5, 9].



$$\Delta G^\circ = A + BT \quad (2)$$

$$\Delta G = \Delta G^\circ + RT \ln K \quad (3)$$

where:

[Z] – element dissolved in liquid steel,

ΔG° – standard Gibbs free energy of forming a non-metallic inclusion with Ce,

T – temperature,

R – gas constant,

K – equilibrium constant of the reaction,

A, B – constants.

Formation reactions can be summarised as shown in Table 1, which presents the reactions and standard Gibbs free energies (ΔG°) for cerium compounds. This indicates that Ce_2S_3 , Ce_3S_4 , and Ce_2O_3 can transform into $\text{Ce}_2\text{O}_2\text{S}$ and CeS, explaining why only these two phases are typically observed in steel. The Gibbs free energy (ΔG) for $\text{Ce}_2\text{O}_2\text{S}$ is significantly less than zero and lower than that of other inclusions, making $\text{Ce}_2\text{O}_2\text{S}$ the most likely phase to form in steel. Table 2 also presents the element interaction coefficients $e(i/j)$ of the components in the steel.

Table 1 (cont.)

Reaction	Formula no.	$\Delta G^\circ = A+B [J \cdot mol^{-1}]$
$[Ce] + 2[O] = CeO_2(s)$	(11)	$-852,720 + 249.96T$
		$-854,274.7 + 249T$
		$-853,600 + 250T$
$2[Ce] + 3[O] = Ce_2O_3(s)$	(12)	$-1,430,200 + 359T$
$Ce_2O_2S(s) + [O] = Ce_2O_3(s) + [S]$	(13)	$-77,500 + 28T$
$[Ce] + [Al] + 3[O] = CeAlO_3(s)$	(14)	$-1,366,460 + 364T$
		$-1,366,460 + 364.3T$
$Ce_2S_3(s) + [Ce] + [S] = Ce_3S_4(s)$	(15)	$-420,541 + 113T$
$[Ce] + 5/2[O] + 1/2[Si] = 1/2Ce_2O_3(s) + SiO_2(s)$	(16)	$-1,106,635 + 286.11T$
$2[Ce] + Al_2O_3(s) = 2[Al] + Ce_2O_3(s)$	(17)	$-1,073,900 + 326T$
$[Ce] + 5/2[O] + 1/2[Si] = 1/2Ce_2O_3(s) + SiO_2(s)$	(18)	$-1,106,635 + 286.11T$
$[Ce] + 1/2[S] + 1/2CaOAl_2O_3(s) = 1/2Ce_2O_2S(s) + [Al] + 1/2[O] + 1/2CaO(s)$	(19)	$-78,057 - 22.24T$
		$-982,839 + 247T$
$Ce_2O_3(s) + [S] = Ce_2O_2S(s) + [O]$	(20)	$77,360 - 28.48T$
		$38,680 - 14.24T$
$[Ce] + Al_2O_3(s) = CeAlO_3(s) + [Al]$	(21)	$-23,900 - 247.30T$
		$-423,900 - 247.30T$
$[Ce] + [N] = CeN(s)$	(22)	$-401,200 + 153.0T$
$[Ce] + 1/2CaOAl_2O_3 = 1/2Ce_2O_3 + 1/2CaO + [Al]$	(23)	$-1,021,604 + 262T$

Table 2

Element interaction coefficients $e(i/j)$ [9, 18]

$e(i/j)$	C	O	S	P	Si	Mn	Al	N	Ce	La
O	-0.45	-0.20	-0.133	0.07	-0.131	-0.021	-3.90	0.057	-12.1	-12.1
S	0.11	-0.27	-0.028	0.029	0.063	-0.026	0.035	0.01	-1.88	-2.79
Ce	0.351	-106	-8.225	1.77	-	0.13	-2.67	-	0.0066	-
Mn	-0.07	-0.083	-0.048	-0.035	-3.0	0	-2.4	-0.091	-0.5	-
Al	0.091	-6.6	0.03	0.033	0.0056	0.035	0.045	-0.053	-6.1	-0.511
Si	0.18	-0.23	0.056	0.11	0.11	0.002	0.058	0.090	-7.7	-
La	-0.279	-105	-12.13	1.734	-	0.28	-2.649	-	-	-0.013

The equilibrium lines for oxide formation lie below those for sulphides, which means that cerium and lanthanum first react with oxygen dissolved in the steel, and then with sulphur (Fig. 2).

The beneficial effect of cerium, due to its high chemical affinity for sulphur and oxygen, is seen in well-deoxidised steel. Cerium oxides have a more negative Gibbs free energy than oxides of manganese, silicon, or aluminium, making them thermodynamically more stable than typical deoxidation products such as Al_2O_3 or SiO_2 . The equilibrium lines for oxide formation lie below those for sulphides, which means that cerium and lanthanum first react with oxygen dissolved in the steel, and then with sulphur. As cerium is consumed during the formation of non-metallic inclusions, its role may shift predominantly to deoxidation and desulphurisation, which reduces its effectiveness as a modifier. At very low Ce concentrations, most cerium reacts to form sulphides, limiting the ability to modify MnS inclusions. With sufficient Mn and S, phases such as Ce_2O_2S , Ce_2S_3 , and MnS may precipitate simultaneously.

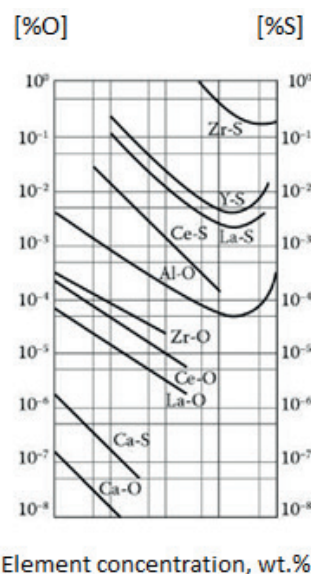


Fig. 2. Equilibrium curves for oxide and sulphide formation at 1,600°C [11]

During aluminium deoxidation, Al_2O_3 forms in the steel, which can react with cerium to produce CeAlO_3 according to Reactions (21) and (14) [14, 19]. The issue of Al_2O_3 inclusion formation as a result of adding Al to steel during refining also applies to the production of high-aluminium TRIP and TWIP steels, where the Al content exceeds 1%, indicating that the contribution of the presented reactions in the transformation of alumina into other forms plays a crucial role. In liquid steel, the formation of non-metallic inclusions such as Ce–P–O and Ce–As–S is also possible, mainly CeP and CeAs. The Gibbs free energy of the formation of these inclusions is higher than that of CeS, Ce_2O_3 , and $\text{Ce}_2\text{O}_2\text{S}$. This means that the occurrence of CeP and CeAs in steel is associated with the segregation of Ce, P, As, and S during solidification and strongly depends on the chemical composition. Their formation limits the unfavourable segregation of P and As at grain boundaries [27].

3. MODIFICATION OF THE CHEMICAL COMPOSITION OF NON-METALLIC INCLUSIONS

The modifying role of cerium is based on reducing and changing existing non-metallic inclusions into complex Ce-containing phases. These modified inclusions become more similar to the steel matrix, so they deform more easily during rolling. Cerium changes Al_2O_3 and MnS inclusions into spherical oxy-sulphide particles, which improves the mechanical properties of the steel. The addition of cerium to liquid steel, MnS and SiO_2 inclusions are transformed into fine $\text{Ce}_2\text{O}_3\cdot\text{SiO}_2$ and $\text{Ce}_2\text{O}_2\text{S}$ particles [6]. A similar effect was also observed for yttrium in alumina inclusions. Research shows that when the amount of cerium increases, MnS inclusions and Al–Si–O systems are converted into phases such as $\text{Ce}_2\text{O}_2\text{S}$, Ce_3S_4 , and CeS. With more cerium, the system becomes dominated by stable Ce–S and Ce–O–S compounds. Yu et al. also reported that after adding Ce, all types of inclusions (MnS, Al_2O_3 , and mixed Al–Si–Ca–O inclusions) eventually transform into CeS and CeO + S phases and the evolution path of Al_2O_3 inclusions follows the sequence: $\text{Al}_2\text{O}_3 \rightarrow \text{CeAlO}_3 \rightarrow \text{Ce}_2\text{O}_2\text{S} \rightarrow \text{Ce}_2\text{O}_2\text{S} + \text{CeS}$. They also identified the presence of the $\text{CeAl}_{11}\text{O}_{18}$ phase in austenitic stainless steel after cerium addition and the CeAlO_3 phase [28].

Calcium is a typical modifier of Al_2O_3 inclusions; however, CaS is formed due to its high chemical affinity for oxygen and sulphur. This phase has a detrimental effect on the steel's properties, as it promotes pitting corrosion of products. Therefore, the degree of calcium treatment must be carefully controlled. In systems containing inclusions of $\text{CaO}\cdot\text{Al}_2\text{O}_3$, reactions with cerium may proceed as described in Reactions (5), (17), (24):



The addition of Ce leads to the modification of $\text{CaS}\cdot\text{C}_{12}\text{A}_7$ inclusions into $\text{C}_{12}\text{A}_7\cdot\text{Ce}_2\text{O}_2\text{S}$. Due to the preferential formation of $\text{Ce}_2\text{O}_2\text{S}$ inclusions, the dissolved sulphur becomes insufficient to react with calcium to form CaS, resulting in the elimination of CaS inclusions. The evolution path of calcium aluminates follows the sequence $\text{CA}_6\text{--CA}_2\text{--C}_{12}\text{A}_7$. Inclusions such as $\text{Ce}_2\text{O}_2\text{S}$ form most easily, and the effect of the modification caused by the cerium addition is significant.

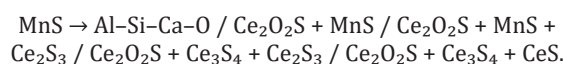
Since cerium is often added in the form of alloys, such as mischmetal, other rare earth metals (REM) also participate in the transformation process. In this case, Al_2O_3 inclusions can transform into more complex phases such as $(\text{Ce,RE})_2\text{O}_3$, $(\text{Ce,RE})_2\text{O}_2\text{S}$, and $(\text{Ce,La})_2\text{O}_2\text{S}$. In aluminium-deoxidised steel, the CeAlO_3 phase forms first and then gradually transforms into Ce_2O_3 and $\text{Ce}_2\text{O}_2\text{S}$.

In steel, inclusions of the $\text{MgO}\cdot\text{Al}_2\text{O}_3$ type also occur. These spinels can likewise be modified by cerium, and the modification path is as follows [28]:

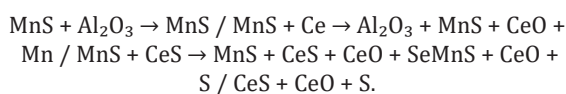


Cerium also modifies MnS inclusions, and this process strongly depends on the contents of O, S, and Ce in the steel. As these elements are gradually consumed and the chemical composition becomes more uniform, sulphur mainly reacts with cerium to form Ce_2S_3 . At the same time, manganese present in the steel combines with the remaining sulphur, which leads to the formation of spherical complex inclusions such as $\text{Ce}_2\text{O}_2\text{S} + \text{Ce}_2\text{S}_3 + \text{MnS}$.

The modification of MnS inclusions in steel was investigated, and the modification process was found to proceed as follows, depending on the Ce content:



Similar results were obtained for pipeline steel. The modification path they proposed is as follows:



MnS modification is also affected by aluminium deoxidation. The Al_2O_3 phase formed earlier can be modified by cerium into CeAlO_3 , and further reactions continue to promote the formation of this phase. This is important for MnS modification because CeAlO_3 particles favour the formation of sulphur-enriched layers around inclusions. As a result, sulphur reacts with cerium to form Ce_2S_3 , while manganese dissolved in the steel can also react with sulphur, which finally leads to the formation of spherical precipitates such as $\text{CeAlO}_3 + \text{Ce}_2\text{S}_3 + \text{MnS}$ [6, 18–19, 29–32].

4. MODIFICATION OF MORPHOLOGY AND SIZE OF NON-METALLIC INCLUSIONS AND STEEL MICROSTRUCTURE

Cerium typically has deoxidising and desulphurising qualities, though it also modifies the grain size. In the case of stainless steel, it improves oxidation resistance. Grain refinement during solidification aims to increase the fraction of equiaxed grains while simultaneously reducing the size of both equiaxed and columnar grains. Depending on the chemical composition of the alloy, steel solidification involves the primary formation of either δ -ferrite (BCC crystal structure) or γ -austenite (FCC crystal structure). During subsequent cooling, δ -ferrite may transform into γ -austenite through a peritectic reaction, whereas the final microstructure can consist of ferrite, pearlite, bainite, or

martensite. Consequently, the solidification pathway and the nature of the phases formed are strongly dependent on the steel grade and chemical composition, implying that different steels require different nucleation substrates and grain-refining mechanisms. Cerium addition results in significant changes in the morphology of the inclusions:

- reduction of their average size,
- transition from irregular to spherical shapes,
- more uniform distribution in the material.

These changes result from the low wettability of cerium inclusions by liquid steel and a high contact angle, promoting their separation and spheroidisation. For instance, inclusions in Ce-modified 55SiCr high-stress spring steel. After adding Ce of 0.02 mass%, the inclusions in the steel mainly consisted of modified MnS, and spherical or ellipsoid Ce–O–S and Ce–S inclusions 1–3 μm in size. However, when the Ce addition was increased to 0.26 mass%, the MnS inclusions disappeared and were replaced by numerous angular and irregular Ce–O and Ce–O–S inclusions over 5 μm in size.

The influence of cerium on the microstructure of steel depends on the steel grade, the type of Ce-based precipitates formed, and the degree of their modification. In most cases, however, its effect is beneficial. In addition to inclusion modification, cerium acts as an effective grain-refining agent. Its addition reduces the size of both equiaxed and columnar grains while increasing the fraction of equiaxed grains, resulting in significant microstructural refinement. Furthermore, cerium suppresses the segregation and precipitation of detrimental elements, thereby improving the mechanical performance of steel.

Microstructural modification and inclusion engineering induced by cerium directly contribute to improved strength, impact toughness, and fatigue resistance of the material. The mechanism of cerium's action is primarily associated with the reduction of stress concentration around non-metallic inclusions and a more homogeneous stress distribution within the steel matrix [29].

Literature data strongly confirm this relationship. The addition of REE reduces the average grain size of austenite in H13 steel from 50 μm to 25 μm , resulting in an increase in impact toughness from 10 J to 19 J. In Nb-microalloyed low-carbon steel a 41.7% reduction in ferrite grain size and a 33.9% decrease in pearlite colony size were observed, accompanied by a 58.7% increase in impact energy. In 2.25Cr–1Mo–0.25V steel the addition of RE elements was reported to lower the bainite transformation start temperature, refine bainitic ferrite, reduce the fraction of carbon-rich martensite–austenite (MA) constituents, and decrease the ductile-to-brittle transition temperature (DBTT). Similarly, the addition of cerium to high-strength aluminium-killed steel reduces the average grain size from 5.06 μm to 2.96 μm and decreases the area fraction of MA constituents from 36.95% to 29.49%. As a result, the absorbed impact energy at -20°C increased from 8.35 J to 19 J [6].

The efficiency of phase nucleation is mainly governed by two factors: a low crystallographic lattice mismatch and a precipitation temperature higher than the steel liquidus temperature. The Ce_2O_3 present acts as nucleation

sites, promoting dendritic refinement and increasing the fraction of equiaxed grains during steel solidification. The addition of rare earth (RE) elements to continuously cast steel (CCS) ingots results in the formation of precipitates with high nucleation potency, which positively influences the ingot microstructure. RE-containing inclusions also improve the impact toughness of cast steel [6]. Cerium is also used in stainless steel to improve oxidation resistance. Its addition reduces the detrimental effect of inclusions by forming stable, spheroidal RE inclusions, thereby improving the mechanical properties of steel. An improvement in the corrosion-resistance of RE-modified stainless steels has also been observed, which is attributed to inclusion modification and the elimination of MnS-type precipitates, thereby reducing the susceptibility to pitting corrosion.

Overall, cerium promotes microstructural refinement, leading to improved mechanical properties, including fatigue strength, weldability, and corrosion resistance. This mechanism is based on heterogeneous nucleation of ferrite or γ -austenite on Ce-containing inclusions such as oxides, sulphides, and oxysulphides. After the addition of cerium to liquid steel, compounds such as Ce_2O_3 , $\text{Ce}_2\text{O}_2\text{S}$, and CeS are formed. Their melting points are higher than the liquidus temperature of steel, allowing them to exist in a solid state during solidification and to act as nucleation sites for new phases. A key factor is the crystallographic lattice mismatch between these particles and ferrite. For Ce_2O_3 , $\text{Ce}_2\text{O}_2\text{S}$, and CeS, it is below 6%, which promotes heterogeneous nucleation. Inclusions such as $\text{Ce}_2\text{O}_2\text{S}$, Ce_2O_3 , CeAlO_3 , Ce_2S_3 , and Ce_3S_4 exhibit lattice misfits below 12% with γ -austenite. Conversely, complex Ce–P and Ce–S inclusions are generally considered detrimental to austenite nucleation. For example, the lattice mismatch of Ce_2S_3 is approximately 1.1%, making it a very effective nucleation site. For CeAlO_3 , it is about 7.7%, indicating moderate effectiveness, whereas for CeP and CeAs, it exceeds 20%, resulting in low nucleation efficiency. The most effective nucleation sites for α - and δ -ferrite are CeS, followed by $\text{Ce}_2\text{O}_2\text{S}$, and then Ce_2O_3 . $\text{Ce}_2\text{O}_2\text{S}$ exhibits the lowest lattice mismatch with α -Fe, making it an effective nucleating agent for α -ferrite. In contrast, the most effective nucleation sites for γ -ferrite are Ce_2O_3 , followed by $\text{Ce}_2\text{O}_2\text{S}$ and CeS. Since α -ferrite is the dominant phase in low-carbon steels, the formation of $\text{Ce}_2\text{O}_2\text{S}$ is crucial for microstructural control. Particles with low lattice mismatch reduce the degree of undercooling required for nucleation. Consequently, the columnar-to-equiaxed transition (CET) occurs earlier, increasing the fraction of equiaxed grains and improving the structural homogeneity. Cerium segregation at dendrite boundaries also restricts grain growth, further refining the microstructure. Moreover, cerium contributes to the elimination of undesirable Widmanstätten structures, which deteriorate the mechanical properties of steel, particularly impacting the toughness [14, 27, 33, 34]. The influence of Ce-containing inclusions on steel performance is also governed by their intrinsic mechanical properties, which may either enhance or deteriorate the overall behaviour of the material. For

example, CeAlO_3 exhibits a higher Vickers hardness and a lower Poisson's ratio than the ferritic matrix, classifying it as a relatively brittle inclusion. However, its coefficient of thermal expansion closely matches that of the surrounding matrix, thereby reducing stress concentration at the inclusion–matrix interface. In contrast, Ce_2O_3 possesses elastic modulus, hardness, and thermal expansion characteristics similar to those of the ferritic matrix, allowing it to behave as a relatively ductile inclusion. $\text{Ce}_2\text{O}_2\text{S}$ exhibits a hardness comparable to that of the matrix and a slightly lower Poisson's ratio; however, its significantly higher coefficient of thermal expansion may promote the development of interfacial stresses during thermal exposure at elevated temperatures [35, 36].

Consequently, doping steel with Ce leads to:

- a reduction in the fraction of columnar grains or their replacement by equiaxed grains in cast steels;
- a similar grain refinement mechanism in ferritic, austenitic, and structural steels, although with variations in intensity;
- a mechanism based on heterogeneous nucleation.

As a result of this process:

- the average grain size is significantly reduced;
- microsegregation of alloying elements is suppressed;
- microstructural homogeneity is improved.

From the perspective of improving the mechanical properties of steel, the formation of an acicular ferrite (AF) microstructure, particularly intragranular acicular ferrite (IAF), is highly desirable. Its randomly oriented interlocking ferrite plates means acicular ferrite not only refines the microstructure and impedes crack propagation but also provides an excellent combination of strength and toughness. Previous studies have demonstrated that inclusions with sizes ranging from approximately 4 μm to 7 μm can act as effective heterogeneous nucleation sites for acicular ferrite formation. The nucleation of acicular ferrite is primarily controlled by the prior austenite grain size and by the characteristics of non-metallic inclusions, including their composition, size, morphology, and population density. Therefore, the deliberate control of inclusion engineering is considered a key strategy for promoting acicular ferrite formation and enhancing room-temperature mechanical properties. In particular, Ce_2O_3 , owing to its low lattice misfit with ferrite, can serve as an effective nucleation substrate for intragranular acicular ferrite during rapid cooling [37].

5. CONCLUSIONS

The role of cerium in modifying non-metallic inclusions, microstructure and properties of steel is determined by its chemical affinity for oxygen and sulphur, as well as by the applied metallurgical and processing conditions. The key effects, benefits, and limitations associated with cerium addition can be summarized as follows:

1. Cerium as an effective inclusion modifier. Cerium exhibits high chemical reactivity towards oxygen and sulphur, enabling effective transformation of non-metallic inclusions (e.g., Al_2O_3 , MnS) into more stable phases such as $\text{Ce}_2\text{O}_2\text{S}$ and CeS.
2. Improvement of mechanical properties. Cerium causes the modification of on metallic precipitates. Inclusion modification leads to a more spherical shape and finer size distribution, reducing stress concentration and improving fatigue resistance, toughness, and overall strength of steel.
3. Influence on microstructure. Cerium promotes grain refinement and increases the fraction of equiaxed grains by acting as nucleation sites, resulting in a more homogeneous steel structure.
4. Optimal cerium content. The most beneficial effects are achieved at cerium contents of approximately 0.01–0.02 wt.%, while higher additions may deteriorate material properties and lead to the formation of unfavourable inclusions.
5. Importance of processing conditions. The effectiveness of cerium depends strongly on the steel composition (especially the O/S ratio), deoxidation practice, and solidification conditions, requiring precise metallurgical control.
6. Limitations of cerium addition. Excessive reactivity of cerium may cause technological issues such as inclusion segregation or clogging during casting.
7. Effect on corrosion and SSC resistance. Modification and reduction of MnS-type inclusions improve corrosion resistance and sulphide stress cracking (SSC) resistance, particularly in aggressive service environments.

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Macroscopic and Microstructural Changes in S355 Steel Welded Joints under Different Fire Scenarios

Marcin Klimek^{a,*} , Mariusz Maślak^a , Marek Stankiewicz^a , Franciszek Hutek^a ,
Krzysztof Pańcikiewicz^b 

^a Cracow University of Technology, Faculty of Civil Engineering, 24 Warszawska St., 31-155 Krakow, Poland

^b AGH University of Krakow, Faculty of Metals Engineering and Industrial Computer Science, 30 A. Mickiewicza Ave., 30-059 Krakow, Poland

*e-mail: marcin.klimek@doktorant.pk.edu.pl

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Abstract

Post-fire assessment of welded steel structures requires knowledge of whether fire exposure may cause permanent changes in the weld and its vicinity. This study examines macro- and microstructural changes in full-penetration butt welded joints made of S355J2+N steel after controlled thermal exposures representative of selected fire-related scenarios. The aim was to determine how maximum temperature, exposure time, and cooling method affect the weld metal, heat-affected zone (HAZ) and base material. Welded joints, prepared in 16 mm plates, were heat-treated at 600°C and 800°C, held for 1 h or 10 h, and cooled freely in a muffle furnace or rapidly in water mist. Etched cross-sections were analysed by optical microscopy. Results showed that short subcritical exposure caused no significant macro- or microstructural changes, whereas prolonged heating below A_{c1} promoted cementite spheroidisation in pearlite and formation of a ferritic–spheroidite structure. Exposure at 800°C, i.e. within the intercritical A_{c1} – A_{c3} temperature range, led to extensive microstructural transformations, with the final microstructure governed mainly by cooling rate. Thus, at 800°C, restructuring of the weld and HAZ becomes more likely, especially after rapid cooling. The originality of the study lies in comparing different fire scenarios for welded joints rather than base steel alone. The findings may support post-fire diagnostics of welded structures and indicate that mechanical testing is needed to relate structural changes to residual load-bearing capacity.

Keywords:

welded joints, S355 steel, post-fire, macrostructure, microstructure

1. INTRODUCTION

Fire is one of the most severe accidental actions to which building structures may be exposed. Due to its unpredictable course, locally variable temperature distribution and the possibility of permanent material degradation, the safety assessment of structures after fire exposure constitutes an important engineering issue.

In steel structures, connections are of particular importance, since their properties largely determine the global reliability of the load-bearing system. This applies especially to welded joints, in which a local, permanent change in the material structure occurs at the manufacturing stage as a result of the high temperature generated during welding, both in the weld metal itself and in the heat-affected zone adjacent to the weld. The impact of a fire episode on a previously executed welded joint should therefore be treated as a subsequent stage of its exposure to elevated temperature, resulting in further microstructural changes and, consequently, in a potential permanent change in material properties. It is therefore essential to identify the processes occurring during welding and during subsequent fire expo-

sure, as they determine the actual degree of degradation of the joint and the possibility of its further safe use [1, 2]. The temperature evolution in a material under fire conditions is inherently random; however, it can generally be stated that these temperatures are lower than those generated in the weld during the welding process.

In the literature concerning the behaviour of welded joints after fire exposure, it is increasingly emphasised that the assessment of mechanical property changes alone is not sufficient to clearly determine the degree of joint degradation. Macrostructural and microstructural analysis is of particular importance, since changes occurring in the weld metal, heat-affected zone and base material determine the subsequent hardness, strength, ductility and fracture resistance. In studies comparing GMAW and FCAW welding processes after fire exposure, Balakrishnan et al. [3] showed that, in S355J2 welded joints, an increase in maximum temperature from 300°C to 900°C leads to gradual microstructural reconstruction, depending on both the temperature and the analysed region of the joint. In the weld metal of GMAW and FCAW joints, the initial microstructure contains different ferrite morphologies, including

acicular ferrite, polygonal ferrite, grain boundary ferrite, Widmanstätten ferrite and bainite. After exposure to 300°C, mainly an increase in grain boundary ferrite and carbide segregation is observed; after 600°C, these processes become more advanced, while after 900°C, the original weld morphology undergoes substantial reconstruction towards a coarser, equiaxed ferritic-pearlitic structure. Ye and Ozaki [4] demonstrated that the microstructure of welded joints after fire exposure strongly depends on the maximum temperature and the cooling method. In the base material and HAZ, furnace cooling promoted the preservation or reconstruction of ferritic-pearlitic structures, whereas water cooling from 800 to 900°C led to the formation of mixed ferritic-bainitic-martensitic structures or fine martensitic structures. Pańcikiewicz et al. [5] also discussed the effect of long-term subcritical annealing on the microstructure of welded joints made of unalloyed structural steel. The authors indicated that, when the exposure time below the A_1 temperature is sufficiently long, the main microstructural process responsible for the observed changes is the spheroidisation of cementite in pearlite. Nasiri and Mirzadeh [6] investigated spheroidisation heat treatment of low-carbon steel and showed that holding at 700°C, i.e. below the estimated A_1 temperature, promotes the formation of spheroidised carbide particles. These observations confirm that prolonged subcritical exposure may lead to significant changes in carbide morphology, which is relevant for interpreting the microstructural changes observed after long-term heating below A_{c1} .

Therefore, understanding the processes occurring both during welding and during subsequent thermal exposure is essential for assessing the post-fire condition of welded steel joints. On this basis, the present study focuses on the macrostructural and microstructural changes in a full-penetration butt weld subjected to controlled thermal exposures representative of selected post-fire conditions. Rather than reproducing the full complexity of an actual fire event, the experimental programme was designed to isolate the effects of maximum temperature, exposure time and cooling conditions on the weld metal, heat-affected zone and base material. The resulting observations are intended to provide a macrostructural and microstructural basis for assessing the condition of welded joints after fire exposure.

2. MATERIALS AND METHODS

2.1. Material and welded joint

The weld joined steel plates with a thickness of 16 mm made of EN 10025-2 S355J2+N steel in the longitudinal direction. The joint was produced using the MAG 135 process in the PA welding position. Welding was carried out as a mechanised, robot-assisted process using a Universal Robot UR10e manipulator and a Merkle HigiPULSE 454 DW welding power source. The joint preparation involved mechanical cutting, machining and grinding. A 1.0 mm diameter Sidergas S7 filler wire, classified according to PN-EN ISO 14341-A as G 46 5 M21 4Si1, was used with M21 shielding gas composed of 82% Ar and 18% CO₂. The weld was made as a multi-run joint with

four runs, using welding currents in the range of 220–260 A, arc voltage of approximately 30.0–32.0 V and welding speed of about 22.7–23.5 cm/min. The corresponding heat input ranged from approximately 1.44 to 1.68 kJ/mm.

In the present study, controlled heat treatment was used to investigate the influence of selected thermal parameters relevant to the post-fire condition of welded joints. The adopted procedure was not intended to reproduce the complete temperature–time history of an actual fire, but rather to provide controlled and repeatable thermal exposure conditions enabling the effects of maximum temperature, exposure time and cooling method to be analysed separately. Particular attention was paid not only to the weld metal itself but also to its immediate surroundings, especially the heat-affected zone and the adjacent base material, where the initial microstructure had already been modified during the welding process.

2.2. Fire scenario simulation

The heat treatment programme was differentiated with respect to three main parameters: maximum heating temperature, holding time at the target temperature, and cooling method. Two temperature levels were selected: 600°C and 800°C. The lower temperature, 600°C, was adopted to represent subcritical exposure, i.e. heating below the austenitic transformation temperature A_{c1} . This scenario was intended to assess whether prolonged exposure below the transformation range may still lead to observable structural changes, such as recovery processes, carbide redistribution or changes in pearlite morphology. The higher temperature, 800°C, was selected as a supercritical exposure level, within the A_{c1} – A_{c3} temperature range, where partial austenitisation occurs and subsequent cooling may significantly influence the final microstructure. Heating within this temperature range promotes the transformation of pearlitic regions into austenite and progressive austenitisation of the surrounding ferritic matrix, followed by renewed transformation during cooling, which explains the pronounced microstructural reconstruction observed after exposure to 800°C. For each temperature level, two holding times were applied: 1 h and 10 h. The two holding times were selected to provide comparative short- and long-duration thermal exposures and to assess the influence of exposure time on the resulting structural changes. In all cases, the specimens were heated at a constant rate of 10°C/min until the target temperature was reached. After the required holding time, the specimens were cooled using one of two different cooling methods. The first method involved free cooling in a muffle furnace, representing slow cooling conditions. In this case, the specimens were left in the muffle furnace for 24 h, allowing them to cool down gradually under furnace conditions. The second method involved rapid cooling in a water mist, intended to reproduce an intensified cooling regime similar to that which may occur during fire-fighting operations. This approach made it possible to compare the influence of both slow and rapid cooling on the macrostructure and microstructure of the weld metal, heat-affected zone and base material after simulated fire exposure. The adopted heat treatment scheme is presented in Figure 1.

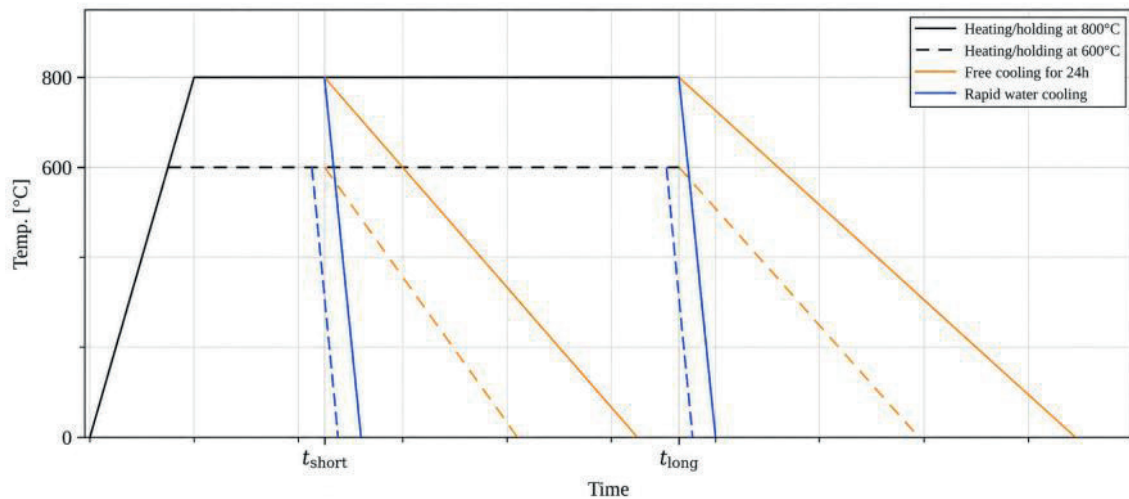


Fig. 1. Temperature–time simulation scheme for the designed fire scenarios

2.3. Equipment used in the study

The heat treatment of the samples in each considered test scenario was carried out in a KJ-M1200-64L-IC muffle furnace, as shown in Figure 2. The macrostructural analysis of the etched sample surfaces was performed using a Leica DM/LM optical microscope (Leica, Wetzlar, Germany), as shown in Figure 3a, with the use of bright field (BF), while the microscopic images presented in this paper were acquired using a Keyence VHX7000N 4K microscope, as shown in Figure 3b.

Before microscopic observations, the prepared cross-sections were etched with Nital in order to reveal the microstructural features. Different regions of the joint, varying in the distance from the heat source and in the initial microstructure formed during the welding process, were subjected to analysis. In particular, the weld metal (WM), the heat-affected zone (HAZ), and the base material (BM) were examined.



Fig. 2. KJ-M1200-64L-IC muffle furnace used for heat treatment of the welded joint samples

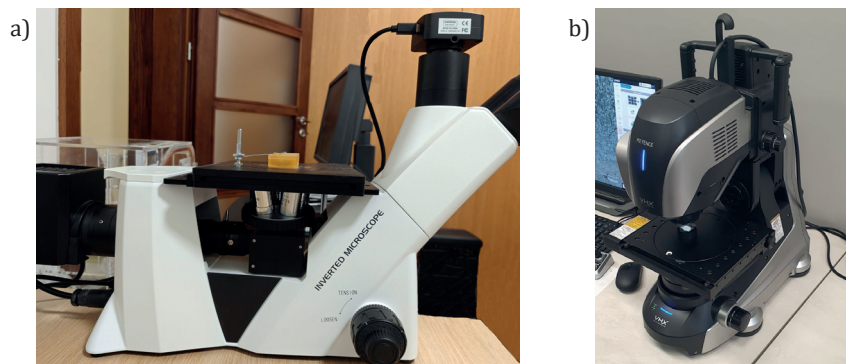


Fig. 3. Optical microscopes used for macrostructural and microstructural observations of the etched sample surfaces: a) Leica DM/LM optical microscope; b) Keyence VHX7000N 4K microscope

3. RESULTS AND DISCUSSION

3.1. Macroscopic observations

Figure 4 presents macrographs of the reference welded joint and the joints exposed to 600°C for 1 h, followed by free cooling or rapid cooling in water, respectively. The reference sample shows a typical macrostructure of a full-penetration butt weld, with a clearly visible weld metal region, heat-affected zones on both sides of the weld, and the adjacent base material. The weld face and root geometry are regular, and no macroscopic welding imperfections are visible in the analysed cross-section. After exposure to 600°C for 1 h, no significant changes in the overall cross-section geometry of the welded joint were observed. Both the weld face and the weld root retained their original shape, and the boundaries between the weld metal, HAZ and base material remained clearly distinguishable.

In the sample subjected to free cooling, the macrostructure is very similar to that of the reference joint, which indicates that short-term subcritical exposure did not cause pronounced macroscopic degradation of the joint. A similar conclusion can be drawn for the sample cooled rapidly in water after exposure to 600°C for 1 h. Although slight differences in etching contrast can be observed within the weld metal and along the HAZ, the general morphology of the joint remains unchanged.

Figure 5 presents macrographs of the reference welded joint and the joints exposed to 600°C for 10 h, followed by free cooling or rapid cooling in water, respectively. After exposure to 600°C for 10 h and free cooling, the general arrangement of the weld metal, heat-affected zone and base material remained clearly visible. However, compared with the reference sample, slight differences in etching contrast can be observed, particularly within the weld metal and along the fusion boundary. The directional structure of the

weld metal is still visible, although the contrast appears slightly more uniform, which may indicate microstructural recovery or partial redistribution of carbide phases during prolonged subcritical exposure. In the sample exposed to 600°C for 10 h and rapidly cooled in water, the macrostructure also remained generally similar to the reference joint.

Figure 6 presents macrographs of the reference welded joint and the joints exposed to 800°C for 1 h, followed by free cooling or rapid cooling in water, respectively. Much more pronounced differences can be observed after exposure to 800°C, not only in comparison with the reference sample but also in relation to all variants exposed to 600°C. In these variants, the macrostructural appearance of the weld metal and the heat-affected zone indicates a more advanced structural reorganisation. This is particularly visible in the regions corresponding to the first and the last weld passes, where the original weld morphology associated with columnar solidification becomes less distinct. The directional solidification pattern observed in the reference joint and still largely preserved after exposure to 600°C is partly blurred after heating to 800°C. This suggests that exposure to 800°C, i.e. above the austenitic transformation temperature A_{c1} , caused significant modification of the initial as-welded structure.

Despite these visible structural changes, the overall geometry of the joint remained stable. The weld face, root and cross-sectional outline did not show macroscopic cracking, discontinuities or significant distortion. Therefore, exposure to 800°C for 1 h did not visibly damage the joint geometry, but it clearly affected the macrostructural image of the weld metal and HAZ. The effect is observed in both cooling variants, although the water-cooled sample shows a slightly sharper and more heterogeneous contrast, which may indicate a stronger influence of the increased cooling rate on the final structure.

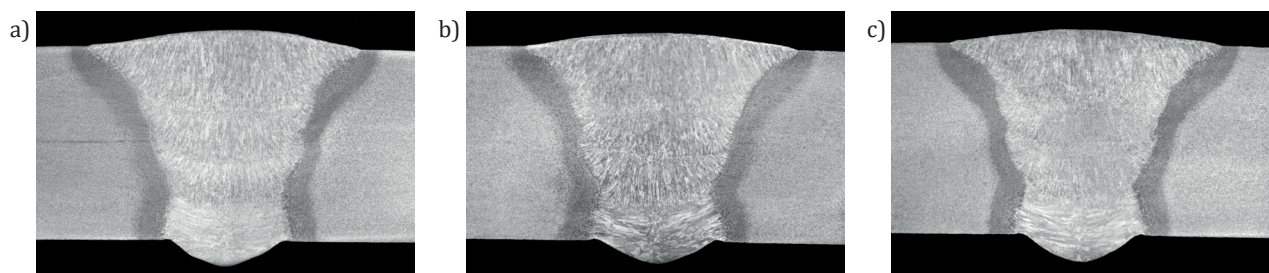


Fig. 4. Macrographs of butt welded joints: a) reference sample, without simulated fire exposure; b) sample exposed to 600°C for 1 h and freely cooled; c) sample exposed to 600°C for 1 h and rapidly cooled in water mist

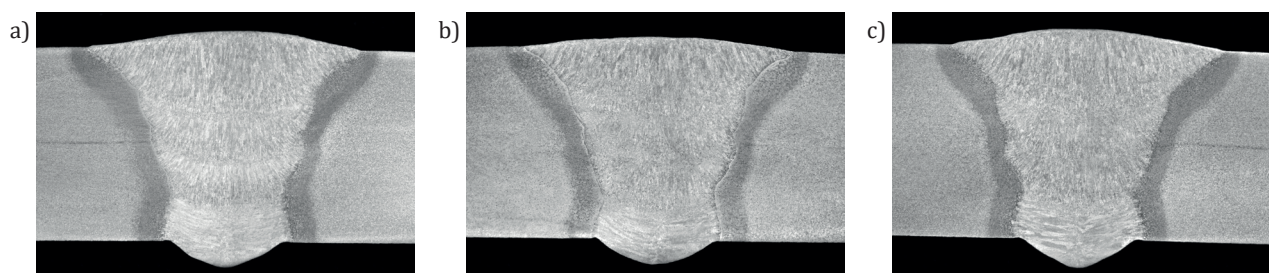


Fig. 5. Macrographs of butt welded joints: a) reference sample, without simulated fire exposure; b) sample exposed to 600°C for 10 h and freely cooled; c) sample exposed to 600°C for 10 h and rapidly cooled in water mist

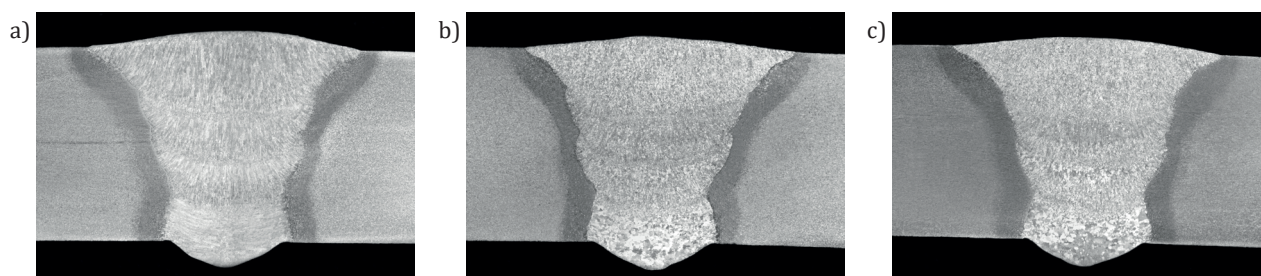


Fig. 6. Macrographs of butt welded joints: a) reference sample, without simulated fire exposure; b) sample exposed to 800°C for 1 h and freely cooled; c) sample exposed to 800°C for 1 h and rapidly cooled in water mist

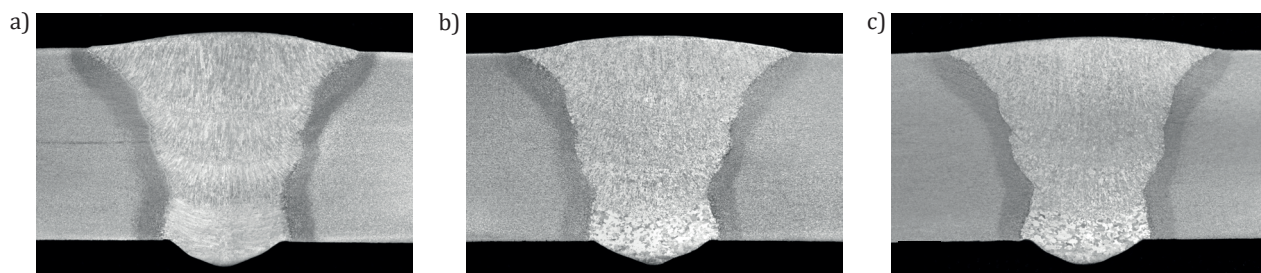


Fig. 7. Macrographs of butt welded joints: a) reference sample, without simulated fire exposure; b) sample exposed to 800°C for 10 h and freely cooled; c) sample exposed to 800°C for 10 h and rapidly cooled in water mist

Figure 7 presents macrographs of the reference welded joint and the joints exposed to 800°C for 10 h, followed by free cooling or rapid cooling in water, respectively. Compared with the reference sample, both heat-treated specimens show a clearly modified macrostructural appearance of the weld metal and the heat-affected zone, analogous to the variant exposed to 800°C for 1 h. The original directional weld morphology associated with columnar solidification is partly blurred, especially in the regions corresponding to the first and last weld passes. This confirms that exposure to 800°C, i.e. above the austenitic transformation temperature A_{c1} , leads to a noticeable reorganisation of the initial as-welded structure. At the same time, comparison with the specimens exposed to 800°C for 1 h does not indicate substantial additional macrostructural changes resulting from the longer holding time. The overall image of the weld metal and HAZ remains generally similar, which suggests that, at this temperature level, the maximum temperature and subsequent cooling conditions are more decisive than the duration of exposure.

The macroscopic observations showed that none of the analysed fire scenarios caused visible damage to the overall geometry of the butt welded joints. The weld face, root and cross-sectional outline remained stable, and no macroscopic cracks, discontinuities or significant distortions were observed. In the variants exposed to 600°C, both after 1 h and 10 h of holding, the macrostructure remained generally similar to that of the reference sample, with only slight differences in etching contrast. This indicates that subcritical exposure did not lead to pronounced macroscopic degradation of the joint. More evident changes were observed after exposure to 800°C. In these

samples, the original directional weld morphology associated with columnar solidification became partly blurred, especially in the regions corresponding to the first and last weld passes. The effect was observed for both holding times, which suggests that, at this temperature level, the maximum temperature and cooling method had a greater influence on the macrostructural appearance than the duration of exposure. Therefore, the macrographs indicate that heating in the vicinity of A_{c3} significantly modifies the internal structural image of the weld metal and HAZ, although without causing visible geometrical damage to the welded joint.

3.2. Microstructural observations

The microstructural analysis was carried out in selected regions of the welded joint, including the base metal, intercritical HAZ, fine-grain HAZ, coarse-grain HAZ and weld metal. The reference sample represents a typical full-penetration butt welded joint in the as-welded condition. A ferritic-pearlitic structure is clearly visible in the base metal, while the individual HAZ subregions show different degrees of structural transformation caused by the welding thermal cycle. The directional character of the structure formed during solidification is preserved in the weld metal.

Figure 8 presents a comparison of the reference sample and the samples exposed to 600°C for 1 h, followed by free cooling and rapid cooling in water mist. After this short subcritical exposure, the general microstructural appearance of all analysed regions remains similar to that observed in the reference joint. The ferritic-pearlitic morphology is still visible in the base metal, and the

directional structure of the weld metal is retained. Only limited changes in contrast and local morphology can be observed in the HAZ and weld metal.

The differences between the freely cooled and water-cooled samples are minor. This is because exposure to 600°C

does not exceed the austenitic transformation temperature A_{c1} ; therefore, rapid cooling does not result in hardening transformations. It can thus be concluded that short-term subcritical exposure causes only limited microstructural changes in the welded joint.

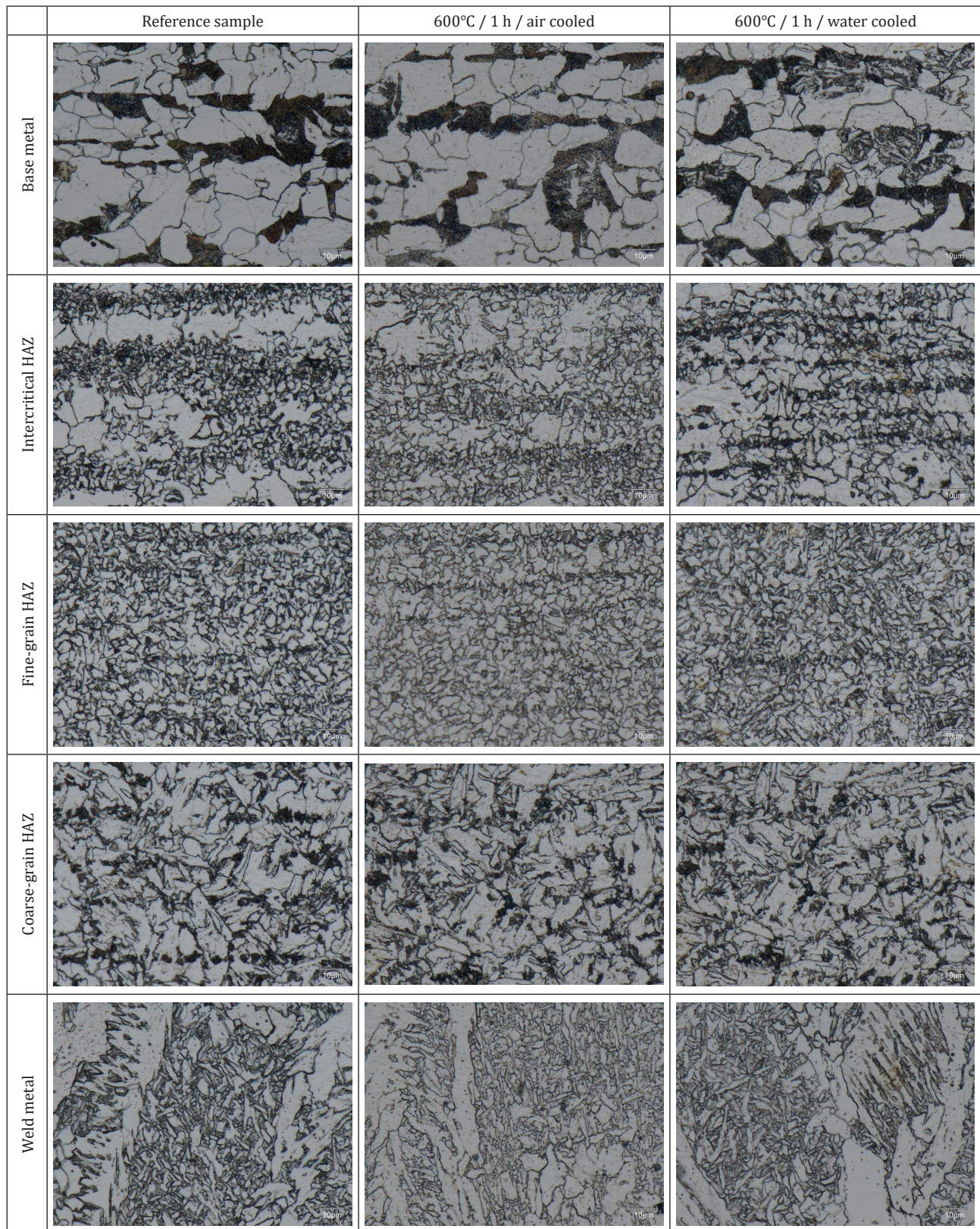


Fig. 8. Microstructural comparison of selected regions of the butt-welded joint in the reference condition and after exposure to 600°C for 1 h followed by free cooling and water-mist cooling

Figure 9 presents the microstructural comparison of selected regions of the butt-welded joint in the reference condition and after exposure to 600°C for 10 h, followed by free cooling or cooling in water mist. The microstructural changes are more pronounced in comparison with

the 1 h exposure experiment, although the temperature still remains below the austenitic transformation temperature A_{c1} . This indicates that, under subcritical conditions, the exposure time may play an important role in the development of structural changes.

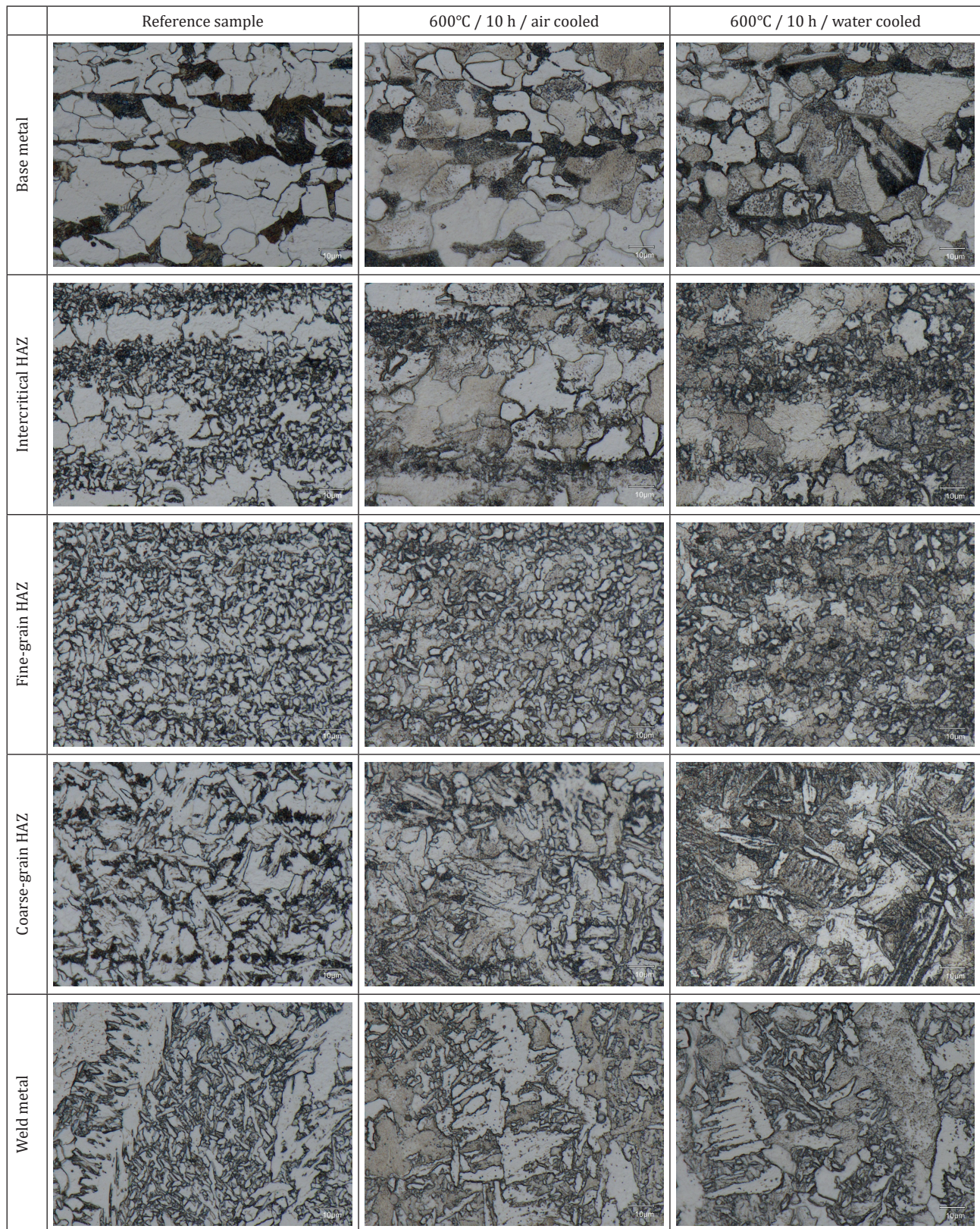


Fig. 9. Microstructural comparison of selected regions of the butt-welded joint in the reference condition and after exposure to 600°C for 10 h followed by free cooling and water-mist cooling

The general ferritic–pearlitic arrangement is still preserved in the base metal. However, compared with the reference sample, the pearlite regions appear less compact and show more evident signs of changes. Similar changes can also be observed in the HAZ subregions, where the morphology becomes less clearly defined and the contrast of dark constituents is more dispersed. Since the exposure temperature did not exceed A_{c1} , these changes should not be interpreted as resulting from austenitisation, but rather as the effect of prolonged subcritical heating, promoting pearlite changes and carbide redistribution.

The directional character of the as-welded structure is still visible in the weld metal, although it is less distinct than in the reference sample. The darker constituents become more fragmented and locally more point-like, which is associated with the spheroidisation of cementite. The differences between free cooling and rapid cooling in water mist are still relatively limited, because rapid cooling from 600°C does not produce hardening transformations. Therefore, the observed changes are mainly related to the prolonged holding time at 600°C rather than to the cooling method itself.

Figure 10 presents the microstructural comparison of selected regions of the butt-welded joint in the reference condition and after exposure to 800°C for 1 h, followed by free cooling or water-mist cooling. At this temperature level, i.e. within the intercritical A_{c1} – A_{c3} temperature range, distinct microstructural changes are visible in all analysed regions of the joint.

In the base metal after exposure to 800°C and free cooling, the ferritic–pearlitic character of the structure is still preserved, but the morphology is clearly modified. Larger and more regular ferrite grains can be observed, while the dark constituents are concentrated in narrower bands. This may indicate that part of the pearlite and carbon-enriched regions transformed into austenite during heating to 800°C. The austenite decomposed again into a ferritic–pearlitic structure during subsequent free cooling. Therefore, the observed structure should not be interpreted as hardened, but rather as a normalised or annealed-like structure.

In the water-cooled base metal, the dark constituents are more irregular and more strongly dispersed. Since 800°C is not high enough to fully homogenise the chemical composition, the former pearlite bands, locally enriched with carbon, could partially transform into non-equilibrium products during rapid cooling. Therefore, the dark regions observed in this variant do not necessarily correspond to typical pearlite; they may include fine products of supercooled austenite transformation, locally including martensitic or bainitic constituents.

In the heat-affected zone, exposure to 800°C caused a strong reconstruction of the initial welded microstructure. A fine polygonal structure is visible in the intercritical HAZ after free cooling. Especially fine grains occur in the regions corresponding to former pearlite, where a higher number of nucleation sites is expected due to the presence of ferrite/cementite interfaces and locally higher carbon content. As a result, this region undergoes relatively effective grain refinement after reheating and cooling.

A different behaviour can be observed in the coarse-grain HAZ. This region had already been characterised by coarser

grains and a more plate-like morphology after welding, as a result of exposure to higher peak temperatures during the welding thermal cycle. After reheating to 800°C, no similarly effective refinement is observed, probably due to the lower number of nucleation sites and partial inheritance of the previous coarse-grained morphology. After cooling in water mist, the HAZ microstructure becomes more heterogeneous, and fine, irregular transformation products are visible, suggesting that rapid cooling prevented the full restoration of a classical ferritic–pearlitic structure.

The most pronounced changes are observed in the weld metal. After exposure to 800°C, both after free cooling and water-mist cooling, the original dendritic or directional weld morphology with a significant amount of fine ferritic constituents is largely lost. Instead, the structure consists mainly of larger polygonal ferrite grains with fine dark particles or precipitates. This indicates that the weld metal was strongly modified by the applied heat treatment. Heating to 800°C caused partial or substantial austenitisation, followed by transformation during cooling into a more equilibrium structure in the freely cooled sample and a more heterogeneous structure in the water-cooled sample. Overall, the observations confirm that the final microstructure is governed mainly by the cooling method and by the initial local microstructure of each welded joint region after exposure to 800°C for 1 h.

Figure 11 presents the microstructural comparison of selected regions of the butt-welded joint in the reference condition and after exposure to 800°C for 10 h, followed by free cooling or cooling in water mist. The observed microstructures are generally similar to those obtained after exposure to 800°C for 1 h, which indicates that, at this temperature level, the maximum temperature and cooling method have a greater influence on the final microstructure than the holding time. At this temperature, the cooling method appears to be the dominant factor controlling the final structural condition of the joint.

The microstructural observations showed that short subcritical exposure at 600°C for 1 h caused only limited changes in the welded joint. The base metal retained its ferritic–pearlitic structure, while the weld metal preserved the directional morphology formed during welding. Extending the exposure time at 600°C to 10 h resulted in more visible changes, mainly related to pearlite changes, carbide redistribution and cementite spheroidisation. However, since this temperature remained below A_{c1} , the differences between free cooling and water-mist cooling were relatively small.

A much stronger microstructural reconstruction was observed after exposure to 800°C. At this temperature, partial or substantial austenitisation occurred, especially in the weld metal and HAZ, leading to a significant modification of the initial as-welded structure. The cooling method had a greater influence on the final microstructure than the holding time in the 800°C variants. Free cooling promoted more equilibrium ferritic–pearlitic or polygonal ferritic structures, whereas water-mist cooling produced a more heterogeneous morphology with possible non-equilibrium transformation products. Overall, the results confirm that the post-fire microstructure of welded joints is governed by the combined effect of the peak temperature, exposure time, and cooling conditions,

with the strongest changes occurring after heating within the intercritical Ac_1 – Ac_3 temperature range. It should be noted that the present microstructural assessment is qualitative. Quantitative determination of grain size and phase or microconstituent fractions was beyond the scope of this study. In particular, pearlite is a microconstituent composed of ferrite

and cementite rather than a separate phase, and its reliable quantitative assessment would require additional dedicated methods beyond conventional etched optical microscopy. Therefore, the present analysis focuses on morphological changes and the identification of the dominant transformation mechanisms.

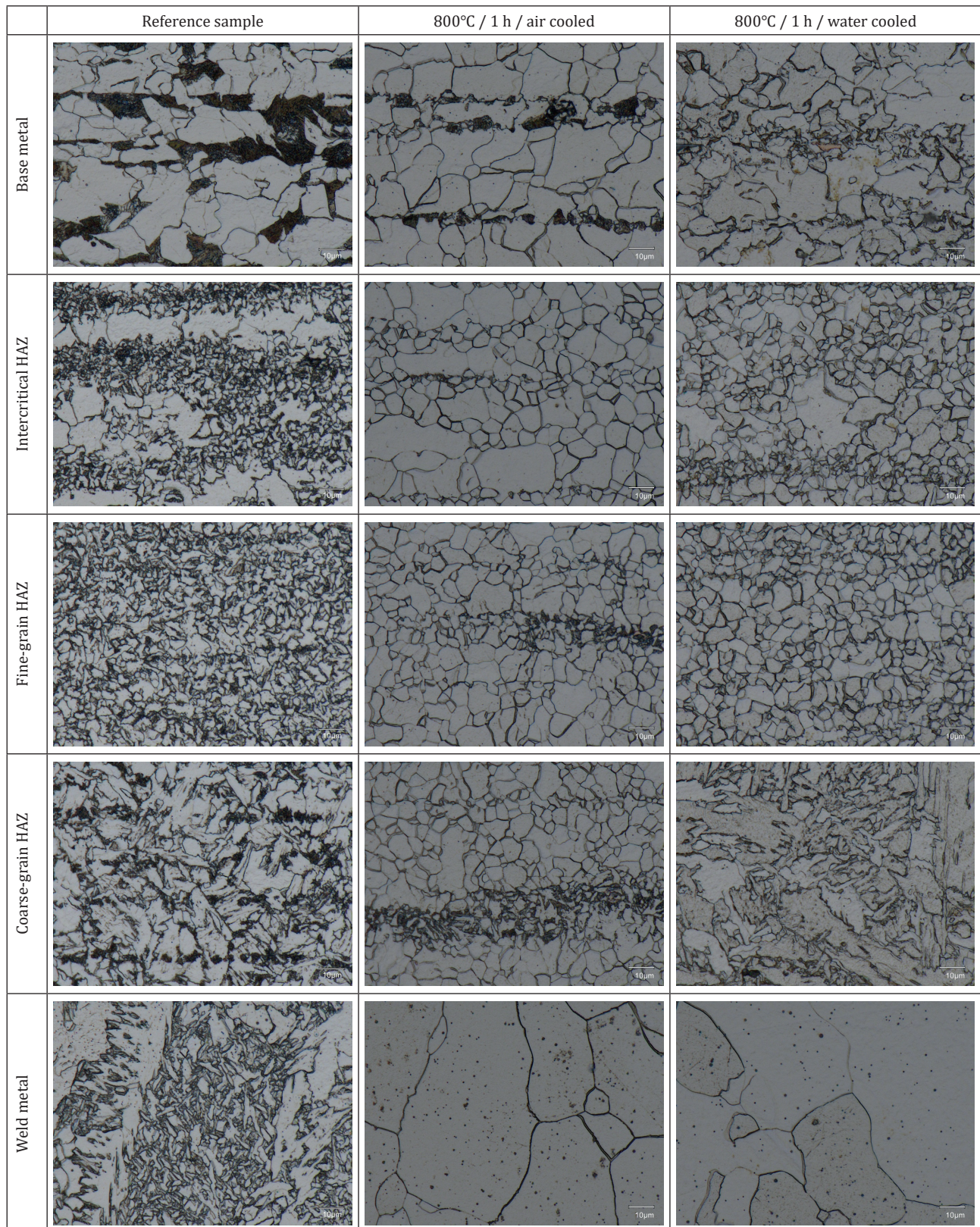


Fig. 10. Microstructural comparison of selected regions of the butt-welded joint in the reference condition and after exposure to 800°C for 1 h followed by free cooling and water-mist cooling

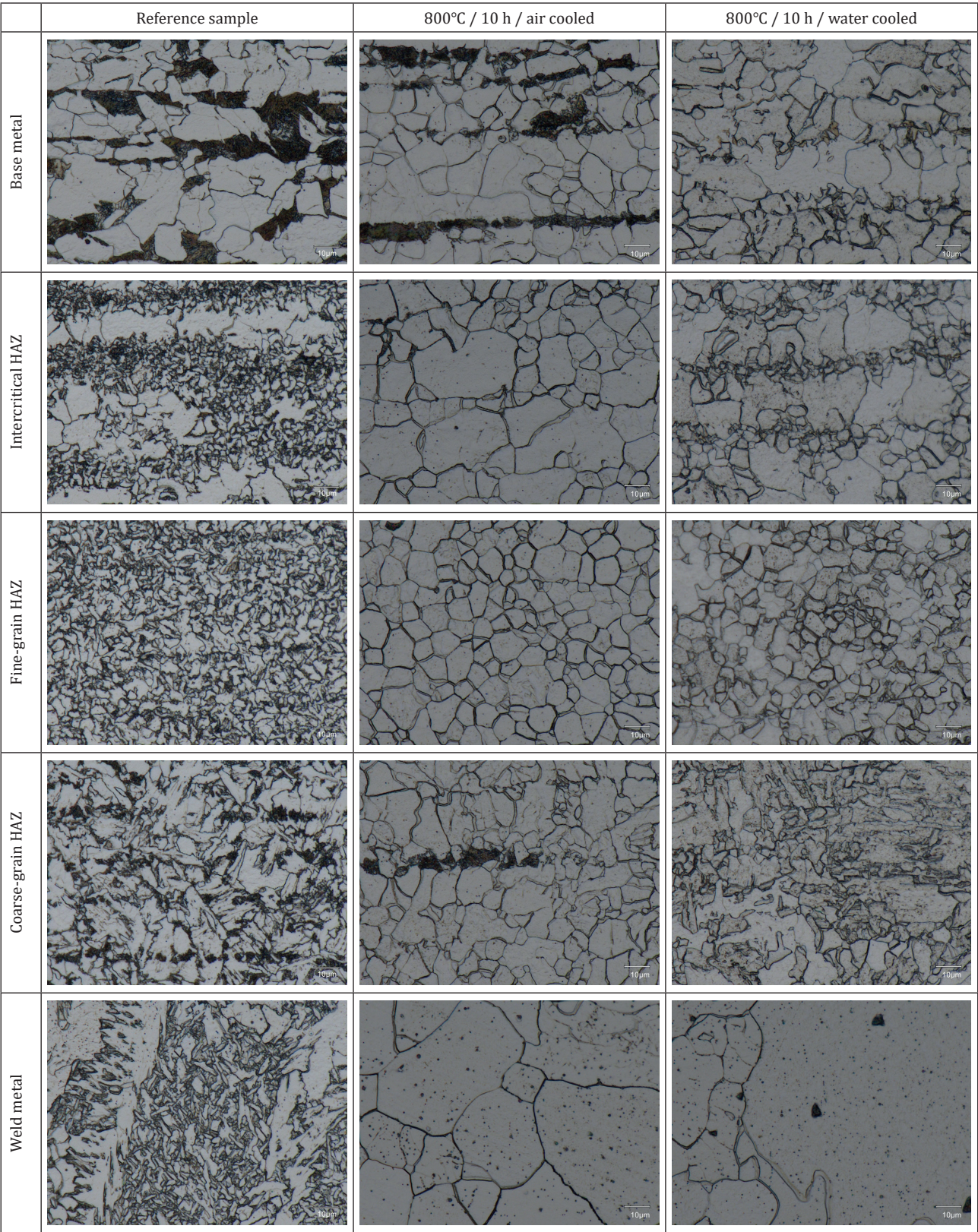


Fig. 11. Microstructural comparison of selected regions of the butt-welded joint in the reference condition and after exposure to 800°C for 10 h followed by free cooling and water-mist cooling

4. CONCLUSIONS AND SUMMARY

The conducted analysis showed that the influence of simulated fire scenarios on the macrostructure and microstruc-

ture of S355J2+N welded joints depends not only on the maximum exposure temperature but also on the holding time and cooling conditions.

The following conclusions can be drawn.

- Short-term subcritical exposure at 600°C for 1 h did not cause significant microstructural reconstruction in the base material, weld metal or heat-affected zone. The overall macrostructural image of the welded joint remained similar to that of the reference sample.
- Extending the exposure time at 600°C to 10 h resulted in more noticeable microstructural changes, mainly associated with pearlite degradation and cementite spheroidisation. This indicates that even subcritical exposure may lead to permanent structural changes if the holding time is sufficiently long.
- Exposure at 800°C, i.e. within the intercritical Ac_1 – Ac_3 temperature range, led to clear reconstruction of the initial as-welded microstructure. These changes were particularly visible in the weld metal and heat-affected zone, where the original directional weld morphology became less distinct.
- The cooling method after exposure at 800°C had a significant influence on the final structural condition of the joint. Free cooling in the furnace promoted the formation of more equilibrium ferritic–pearlitic structures, whereas rapid cooling in water mist may favour the formation of non-equilibrium transformation products.
- The post-fire assessment of welded joints should therefore not be based solely on the maximum temperature reached during fire exposure. The holding time and cooling conditions should also be considered, as these parameters determine the degree of permanent microstructural changes and may significantly affect the residual mechanical properties of the welded joint.

Further complementary studies are required to verify and extend the present conclusions, particularly by determining the mechanical properties of the welded joints after simulated fire exposure. Such tests should include hardness measurements, tensile testing, and impact toughness assessments in order to directly relate the observed macrostructural and microstructural changes to the residual mechanical performance of the joint.

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