Effect of Shielding Gas on Microstructure and Mechanical Properties of 1000 MPa Grade Deposited Metal (Postprint)

Authors: An Tongbang, Tian Zhiling, Shan Jiguo, Wei Jinshan

Date: 2023-03-19T00:00:00+00:00

Abstract

The effects of different shielding gases (Ar+5%CO2, Ar+10%CO2, Ar+20%CO2, and Ar+30%CO2) on the microstructure and strength-toughness of 1000 MPa grade high-strength deposited metal were investigated. The results show that when the CO2 content is 20%, the deposited metal exhibits optimal mechanical strength and toughness, with a yield strength of 980 MPa, room-temperature impact energy of 72.6 J, and -40 ℃ impact energy of 52 J. Microstructural observation and analysis results indicate that with increasing CO2 content in the shielding gas, the bainite lath content in the deposited metal microstructure increases, and the distribution morphology of bainite laths transforms from parallel to interwoven. The interwoven bainite laths partition and refine prior austenite grains, thereby refining martensite laths. The bainite content and the distribution morphology of martensite/bainite laths are the fundamental factors determining the mechanical properties of the deposited metal. Higher bainite content is not always beneficial; an optimal proportion exists. With further increase in CO2 content in the shielding gas, the number of inclusions in the deposited metal increases, their size enlarges, and the content of main components changes. When the CO2 content in the shielding gas reaches 30%, larger-sized inclusions appear, leading to reduced toughness of the deposited metal.

Full Text

Effect of Shielding Gas on Microstructure and Mechanical Properties of 1000 MPa Grade Deposited Metals

AN Tongbang1),2)**, TIAN Zhiling**2)**, SHAN Jiguo**1)**, WEI Jinshan**2)
1) Department of Mechanical Engineering, Tsinghua University, Beijing 100084
2) Central Iron & Steel Research Institute, Beijing 100081

Correspondent: TIAN Zhiling, professor, Tel: (010)62181049, E-mail: tianzhl@cisri.com.cn

Manuscript received: 2015-06-02, **in revised form:** 2015-09-06

Abstract

This study investigates the influence of different shielding gases (Ar+5%CO_2, Ar+10%CO_2, Ar+20%CO_2, and Ar+30%CO_2) on the microstructure and strength-toughness properties of 1000 MPa grade high-strength deposited metals. The results demonstrate that when the CO_2 content reaches 20%, the deposited metal exhibits optimal mechanical properties, with a yield strength of 980 MPa, room-temperature impact energy of 72.6 J, and -40 ℃ impact energy of 52 J. Microstructural observation and analysis reveal that as the CO_2 content in the shielding gas increases, the bainite lath content in the deposited metal microstructure increases, and the bainite lath distribution transforms from parallel to interwoven configurations. The interwoven bainite laths partition and refine prior austenite grains, thereby refining martensite laths. The bainite content and the distribution morphology of martensite/bainite laths constitute the fundamental determinant of mechanical properties. However, higher bainite content is not necessarily better, as an optimal proportion exists. With further increases in CO_2 content, the quantity and size of inclusions in the deposited metal increase, and the composition of major constituents changes. When the CO_2 content reaches 30%, large-sized inclusions appear, leading to reduced toughness in the deposited metal.

Keywords: 1000 MPa grade deposited metal, shielding gas, strength and toughness, martensite/bainite lath, distributed morphology, oxide inclusion

1. Introduction

Gas metal arc welding offers advantages including low cost, high quality, and wide applicability, making it extensively used in construction machinery, shipbuilding, and offshore engineering industries. Weld performance depends on both welding materials and processes, with different shielding gases exerting varying effects on weld microstructure and properties. Research has shown that shielding gas composition influences weld oxygen content, inclusion characteristics, and distribution states, thereby affecting the transformation behavior of supercooled austenite in welds. Seo et al. [2] demonstrated that for 600 MPa grade deposited metals, oxygen content increases with rising CO_2 content in Ar+CO_2 shielding gas. At low oxygen levels (below 100$×10{-6}), themicrostructureconsistsofbainite; atmediumoxygenlevels(approximately150×10{-6}), thestructurebecomesnearlyentirelyacicularferritewithoptimalimpactproperties; whileathighoxygenleve
6}), thestructurecomprisesbothacicularferriteandbainite.Yuetal.[3]reportedthatultra−lowcarbonbainiticdepositedmetalsconsistofbainitelathsandminoracicularferrite, withoptimalperformance
1.5{2}$ *shielding; when using Ar+4%*CO_2*, carbon pickup occurs with carbide*

formation and numerous oxide inclusions, both causing sharp toughness deterioration. Studies on 580–780 MPa grade weld metals [4-6] indicate that impact toughness improves with decreasing weld oxygen content, though other research [7-9] suggests that at 200$×10^{-6}$ oxygen content, oxide inclusions most effectively promote acicular ferrite formation, yielding optimal acicular ferrite content and maximum toughness. Terashima and Hart [10] investigated shielding gas effects on low-strength (624 MPa) and medium-strength (688 MPa) welds, finding that at 200$×10^{-6}$ oxygen content, the microstructure was primarily acicular ferrite with optimal strength-toughness balance; for high-strength (778 MPa) welds, oxide inclusions in the martensitic matrix acted as crack initiation sites, reducing toughness when oxygen content was below 140$×10^{-6}, whiletoughnessimprovedwhenoxygencontentdecreasedto20×10^{-6}.Goudaetal.[11]studied950MPagradeweldmetals, concludingthatincreasingCO{2}$ content in Ar+CO_2 shielding gas decreased both weld hardness and impact toughness, with the microstructure transforming from lath martensite and carbide-free bainite to bainite-dominated structures containing granular bainite with martensite/austenite (M/A) constituents, both of which increased with CO_2 content and degraded weld performance.

These studies demonstrate that shielding gas primarily determines weld oxygen content and oxide inclusion characteristics, subsequently influencing weld microstructure and properties. For low-strength welds (tensile strength σb $\leq$ 580 MPa), acicular ferrite represents the optimal microstructure for strength-toughness balance, with the best acicular ferrite content obtained at approximately 200$×10^{-6}$ oxygen content, where inclusion characteristics critically determine acicular ferrite formation [12,13]. For medium-strength welds (580–780 MPa), satisfactory strength-toughness combinations can also be achieved at lower or suitable oxygen levels. However, research on shielding gas effects on 1000 MPa grade high-strength deposited metal microstructures and properties remains scarce. Therefore, investigating and elucidating these effects holds significant importance for developing high-strength welding consumables and practical applications.

This work employs four different Ar+CO_2 shielding gas mixtures to prepare deposited metals through metal active gas (MAG) welding, focusing on the influence of shielding gas on the microstructure and properties of 1000 MPa grade low-alloy high-strength deposited metals. The study explores the formation mechanisms of deposited metal microstructures and the primary factors affecting their strength and toughness, aiming to provide guidance for the application of 1000 MPa grade welding materials.

2. Experimental Methods

Welding test plates were fabricated from quenched-and-tempered Q960 low-alloy high-strength steel with a chemical composition (mass fraction, %): C 0.14, Si 0.41, Mn 1.18, (Ni + Cr + Mo) 0.89, Ti 0.01, Cu 0.015, Nb 0.014, Fe balance. Sample dimensions were 260 mm × 200 mm × 20 mm. The welding consumable was a self-developed 1000 MPa grade Mn-Ni-Cr-Mo gas-shielded solid wire with a diameter of 1.2 mm and chemical composition: C 0.10, Si 0.59, Mn 1.83, (Ni + Cr + Mo) 3.91, Ti 0.091, S 0.0047, P 0.0062, Fe balance. Welding was performed using a YH-751 automatic gas-shielded welding machine at a constant heat input of 13 kJ/cm, interpass temperature of 110 ℃, and gas flow rate of 20 L/min. Shielding gases were: 95% Ar + 5% CO_2, 90% Ar + 10% CO_2, 80% Ar + 20% CO_2, and 70% Ar + 30% CO_2 (volume fractions). The chemical compositions of deposited metals are presented in Table 1 .

Metallographic specimens were sectioned from the final weld bead, ground with abrasive paper, and polished. Microstructures were revealed using 3% nitric acid alcohol solution, while color metallography employed Lepera etchant (1:1 mixture of 1% $Na_2S_2O_5$ aqueous solution and 4% picric acid alcohol solution, etched for 3 min) [14,15]. A MEF4-M optical microscope (OM) was used for microstructural and color metallographic observations, with SISC IAS8.0 software employed for color metallography and inclusion statistical analysis. An S-4300 cold field-emission scanning electron microscope (SEM) was utilized for fracture surface examination, equipped with an EDAX Genesis 6.0 energy-dispersive spectrometer for inclusion composition analysis. HV-5 Vickers hardness tester measured the hardness of final weld beads under a 49 N load for 10 s. Microstructural observation was conducted on an H-800 transmission electron microscope (TEM) at 175 kV. TEM specimens were prepared by grinding to less than 50 m thickness, followed by electropolishing using an MTP-1A magnetic-driven twin-jet apparatus at -25 to -20 ℃ with 6% perchloric acid ethanol solution at 25 V and 60 mA. An Oxford Nordlys F+ field-emission SEM performed electron backscatter diffraction (EBSD) area scans over 200 m × 200 m regions with a 0.1 m step size, with data analyzed using HKL-Channel 5 software. Tensile and impact test specimens were prepared according to national standard GB/T 8110-2008. Impact specimens measured 10 mm × 10 mm × 55 mm with a 2 mm deep V-notch at 45° angle.

2.1 Chemical Composition of Deposited Metals

Table 1 reveals that the primary compositional differences among deposited metals are that oxygen content increases with rising CO_2 content in shielding gas, nitrogen content shows little variation, and all other elements exhibit decreasing trends. To stabilize the arc, shielding gases contain certain amounts of oxidizing gas, making metal oxidation and oxygen pickup inevitable during welding. Higher CO_2 content in shielding gas leads to higher oxygen content and lower contents of deoxidizing elements Si, Mn, and Ti in deposited metal. With Ar+30%CO_2 shielding gas, deposited metal exhibits the highest oxygen content and lowest Si, Mn, and Ti contents.

2.2 Mechanical Properties of Deposited Metals

Figure 1 [Figure 1: see original paper] illustrates the relationship between mechanical properties and CO_2 content in shielding gas. As shown in Figure 1a, deposited metal strength decreases with increasing CO_2 content. When CO_2 content rises from 5% to 30%, tensile strength decreases from 1172 MPa to 1044 MPa, and yield strength decreases from 1038 MPa to 921 MPa, while elongation and reduction of area remain essentially unchanged. Figure 1b demonstrates that impact toughness initially increases then decreases with rising CO_2 content. At 20% CO_2 content, deposited metal achieves peak yield strength of 980 MPa with a yield ratio of 0.84, along with maximum room-temperature and -40 ℃ impact energies of 72.6 J and 52 J, respectively. These results indicate that shielding gas not only affects deposited metal strength but also serves as a critical factor controlling toughness. Moreover, lower CO_2 content is not necessarily better; an optimal shielding gas composition exists for strength-toughness balance. Based on these results, 80%Ar+20%CO_2 yields the best combination of strength and toughness.

2.3 Microstructural Characteristics of Deposited Metals

Figure 2 [Figure 2: see original paper] presents OM images of final weld bead microstructures for different shielding gases, while Figure 3 [Figure 3: see original paper] shows TEM images and residual austenite electron diffraction analysis. Vickers hardness measurements of final weld beads reveal average HV5 values of 404.25, 381.47, 366.71, and 358.39 HV with increasing CO_2 content. Based on Figures 2, 3, and hardness results, all deposited metals consist of martensite/bainite lath mixed structures with minor interlath residual austenite. Figure 2a shows the microstructure with Ar+5%CO_2, where columnar grains are primarily composed of martensite lath packets sharing the same habit plane, with relatively low bainite content. As CO_2 content increases, bainite quantity rises and its distribution becomes more diversified, with interwoven bainite laths partitioning and refining the intragranular structure (Figure 2b). With Ar+20%CO_2 (Figure 2c), the microstructure comprises uniformly distributed martensite blocks and bainite laths, where interwoven bainite laths further refine the martensite blocks. At Ar+30%CO_2 (Figure 2d), bainite laths coarsen and locally form bainite packets, while martensite content relatively decreases.

TEM observation (Figures 3a and b) reveals high dislocation densities in both martensite and bainite laths, with no carbides present in bainite laths (Figure 3b), indicating carbide-free bainite. Bainite formed during continuous cooling can be classified into two types [12]: carbide-free bainite laths generated at high cooling rates, and granular bainite formed at low cooling rates. Granular bainite consists of a ferrite matrix with distributed M/A islands, where M/A constituents are low-ductility phases that typically promote cracking. However, no granular bainite was observed in any of the four deposited metals. Figures 3c and d show thin film/lamellar residual austenite existing between martensite/bainite laths. Residual austenite represents untransformed austenite retained at room temperature during continuous transformation. With its FCC structure and good ductility/toughness, thin film residual austenite between martensite/bainite laths benefits toughness improvement [14].

2.4 Martensite/Bainite Sub-Unit Orientation Distribution

TEM observation indicates that deposited metals primarily consist of martensite and bainite laths, forming a hierarchical sub-structure comprising packets, blocks, and laths. EBSD effectively reveals block structure distribution morphology. Figure 4 [Figure 4: see original paper] shows EBSD orientation maps and inverse pole figures of block substructures for different shielding gases, where different colors represent different orientations: blue for (111), red for (001), and green for (101). Block morphologies exhibit two distribution patterns: interwoven and parallel. Interwoven block distributions effectively inhibit crack propagation, while parallel distributions facilitate it. Additionally, interwoven morphology promotes microstructural refinement and enhances impact toughness.

With Ar+5%CO_2 (Figure 4a), similarly oriented blocks within the same columnar grain distribute parallel to each other, while differently oriented blocks also show predominantly parallel arrangements. As CO_2 content increases, both block size and distribution morphology change significantly, showing a trend toward finer, interwoven blocks with different orientations that favor further refinement. At Ar+20%CO_2 (Figure 4c), the interlocking and interwoven morphology among differently oriented blocks within columnar grains becomes most pronounced, with relatively fine block sizes. While Ar+10%CO_2 and Ar+30%CO_2 deposited metals (Figures 4b and d) also exhibit some interwoven block morphologies, they contain partially larger blocks with less distinct interwoven patterns, and the interwoven distribution is less pronounced than at Ar+20%CO_2. The EBSD-observed morphologies are consistent with the metallographic structures shown in Figure 2.

2.5 Inclusion Characteristics in Deposited Metals

Table 2 summarizes inclusion statistics for the four shielding gas conditions. With increasing CO_2 content, the total number of inclusions shows an upward trend. Figure 5 [Figure 5: see original paper] illustrates inclusion size distributions. Analysis of Table 2 and Figure 5 reveals that the percentages of inclusions smaller than 1 m are 98.87%, 97.61%, 96.21%, and 95.09% for Ar+5%CO_2, Ar+10%CO_2, Ar+20%CO_2, and Ar+30%CO_2, respectively. Maximum inclusion diameters are 1.118, 1.365, 1.408, and 1.447 m, with number densities of $10.0\$\times 10^{\{4\},9.8\times 10}\{4\},12.3\times 10^{\{4\},and12.9\times 10}\{4\}\$$ mm^{-2}. These results demonstrate that inclusions are predominantly small-sized, though their quantity increases and the proportion of fine inclusions decreases with rising CO_2 content. Inclusion characteristics depend directly on welding method and materials. Since all welding conditions except shielding gas were identical, shielding gas can be considered the primary cause of these inclusion variations.

Chemical reactions in the molten pool at high temperatures control weld metal composition, with welding consumables, base metal composition, and welding conditions all influencing the final chemistry. Weld microstructure and properties depend on chemical composition and heat input, with shielding gas being the only variable in this study. Table 1 shows that increasing CO_2 content raises oxygen content while decreasing C, Si, and Mn contents, reducing the carbon equivalent (Ceq) that governs hardenability.

Figures 2 and 3 confirm that the microstructures consist primarily of martensite and bainite laths. Figure 6 [Figure 6: see original paper] presents color metallographic OM images after Lepera etching, where brown represents martensite and dark blue represents bainite [15-18]. Quantitative analysis shows bainite contents of 8%, 15.4%, 23.8%, and 29.6% as CO_2 content increases from 5% to 30%. The increased CO_2 content intensifies oxidation reactions in the arc region, releasing heat equivalent to increased welding heat input and reducing post-weld cooling rate, which promotes bainite formation [19-21]. Additionally, the bainite transformation temperature at 50% volume fraction (B_{50}) and martensite start temperature (Ms) were calculated using the following equations [22]:

$$B_{50} = 830 - 270C - 90Mn - 37Cr - 83Mo - 60Ni$$
$$M_s = 521 - 353C - 22Si - 17.3Mn - 25.8Cr - 24.3Ni - 20Cu - 10Mo$$

where ω represents the mass fraction of each element in the deposited metal.

Figure 7 [Figure 7: see original paper] shows B_{50} and Ms variations with CO_2 content. As CO_2 content increases, reduced alloy element contents enlarge the temperature interval between B_{50} and Ms. A larger B_{50}-Ms temperature difference implies more bainite content because higher bainite transformation start temperature (Bs) and expanded Bs-Ms interval provide longer duration for bainite transformation.

TEM observation reveals that bainite primarily consists of bainite laths, while OM images show characteristic bainite laths growing in different directions from prior austenite columnar grain boundaries, with parallel bainite laths forming bainite blocks. Due to abundant lattice defects at grain boundaries, bainite preferentially nucleates at boundaries and grows inward, forming primary bainite laths (Figure 2a). With increasing CO_2 content, reduced cooling rate and expanded B_{50}-Ms interval facilitate formation of many differently oriented secondary laths on primary bainite laths. Additionally, intragranular bainite laths nucleate and grow on effective inclusions (Figure 8 [Figure 8: see original paper]), forming primary laths that subsequently develop secondary structures on their sides [3,23]. Bainite laths nucleating at both grain boundaries and intragranular sites significantly increase microstructural complexity (Figures 2c and 4c). When temperature decreases to Ms, the partitioned and refined untransformed austenite between bainite laths transforms to martensite, yielding the final interwoven bainite lath and interlath martensite structure.

Research [24,25] indicates that martensite/upper bainite mixed structures degrade properties, while martensite/lower bainite mixtures improve strength-toughness balance because pre-formed lower bainite partitions prior austenite grains, refining subsequent martensite packet sizes. With Ar+5%CO_2, low bainite content provides limited partitioning effect. During deformation, the surrounding higher-strength martensite restricts bainite deformation, creating strain incompatibility that generates high stress concentrations at bainite/martensite interfaces, promoting cracking and rapid crack propagation along interfaces, thus deteriorating toughness. As CO_2 content increases, higher bainite content enhances partitioning effects, reducing martensite packet size and producing more interwoven block morphologies. During fracture, cracks encountering bainite/martensite interfaces or martensite packet boundaries are arrested and deflected. Whether cracks choose new paths through bainite or bypass along interfaces, the crack propagation path lengthens, increasing energy consumption and improving toughness. Figures 2 and 4 show that at 20% CO_2 content, the interlocking morphology among differently oriented blocks is most pronounced, providing optimal partitioning for toughness improvement. Analysis of mechanical properties, microstructure, and morphology indicates that martensite/bainite mixed structures outperform single martensite structures, and that optimal bainite content exists rather than simply more being better.

Inclusions in deposited metals are predominantly spherical. Some bainite laths within columnar grains nucleate on effective spherical inclusions and grow radially (Figure 8). When high interfacial energy and strain energy exist between inclusions and austenite matrix, and the misfit between ferrite laths and inclusions is small, the distortion energy for ferrite nucleation on inclusions decreases, reducing nucleation work and promoting intragranular nucleation [26,27]. No acicular ferrite was observed in this work; effective inclusions served only as nucleation sites for bainitic ferrite. Studies [2] indicate that oxide inclusion nucleation has specific composition and size requirements, with inclusions 0.2–0.6 m in diameter and surface TiO_2 layers being most effective. EDS analysis of randomly selected inclusions (Table 3) shows they primarily consist of Ti, Mn, Si, and Al oxides and sulfides. With increasing CO_2 content, MnO/S and SiO_2 contents increase while TiO_2 content decreases.

At Ar+30%CO_2, elevated oxygen content increases average inclusion diameter, making larger inclusions less effective as nucleation sites. Moreover, for higher-strength welds, smaller inclusions can initiate cleavage, so large inclusions more readily become crack sources at inclusion-matrix interfaces, degrading toughness [28,29]. Figure 9 [Figure 9: see original paper] shows cleavage fracture initiated from inclusions in Ar+30%CO_2 deposited metal at -40 °C, where cracking begins at inclusions under normal stress, forming disc-shaped microcracks that propagate into the matrix.

Analysis of the relationship between shielding gas and mechanical properties indicates that Ar+20%CO_2 produces optimal strength-toughness balance. For 1000 MPa grade high-strength deposited metals, lower CO_2 content is not necessarily better; appropriate CO_2 content facilitates formation of optimally combined microstructures and distributions while controlling oxide inclusion content to prevent large inclusions that damage toughness.

3. Conclusions

1. With increasing CO_2 content in shielding gas, 1000 MPa grade deposited metal strength decreases slightly while impact toughness first increases then decreases. At Ar+20%CO_2, deposited metal achieves optimal strength-toughness balance: yield strength of 980 MPa, room-temperature impact energy of 72.6 J, -40 ℃ impact energy of 52 J, and yield ratio of 0.84.
2. As CO_2 content increases, bainite content in deposited metal increases and bainite lath morphology changes from parallel to interwoven, which partitions and refines prior austenite grains and consequently refines martensite laths. The morphology of martensite/bainite laths and bainite content fundamentally determine mechanical properties. Higher bainite content is not always beneficial, and an optimal proportion exists.
3. Different shielding gases produce variations in inclusion quantity, size, and composition. With increasing CO_2 content, inclusion quantity and size increase while major constituent contents change. At 30% CO_2 content, increased large-sized inclusions deteriorate deposited metal toughness.

References

[1] Liu S, Olson D L. Weld J, 1986; 65(6): S139
[2] Seo J S, Lee C H, Kim H J. ISIJ Int, 2013; 53: 279
[3] Yu S F, Qian B N, Guo X M. Acta Metall Sin, 2005; 41: 1082 (Yu Shaofei, Qian Bainian, Guo Xuming. Acta Metallurgica Sinica, 2005; 41: 1082)
[4] Powell G L F, Herfurth G. Metall Mater Trans, 1998; 29A: 2775
[5] Utterberg B, Svensson L E. Sci Technol Weld Join, 2002; 7: 363
[6] de Rissone N M R, Bott I D, de Vedia L A, Surian E S. Sci Technol Weld Join, 2003; 8: 113
[7] Terashima T, Tsuboi J. Proc Natl Meet JWS' 78, Tokyo: Japan Welding Society, 1978: 28
[8] Terashima H, Tsuboi J. Met Constr, 1982; 14: 648
[9] Terashima H, Hart P H M. Weld J, 1984; 63(6): S173
[10] Terashima S, Bhadeshia H K D H. Sci Technol Weld Join, 2006; 11: 509
[11] Gouda M, Takahashi M, Ikeuchi K. Sci Technol Weld Join, 2005; 10: 369
[12] Sato Y, Hayakawa N, Kuwana. Q J Jpn Weld Soc, 1992; 10: 422
[13] Prokic- Cvetkovic R, Milosavljevic A, Sedmak A, Popovic O. J Serb Chem Soc, 2006; 71: 313
[14] Peng Y, Wang A H, Xiao H J, Tian Z L. Acta Metall Sin, 2012; 48: 1281 (Peng Yun, Wang Aihua, Xiao Hongjun, Tian Zhiling. Acta Metallurgica Sinica, 2012; 48: 1281)
[15] Lepera F S. J Met, 1980; 32(3): 38

[16] Girault E, Jacques P, Harlet P, Mols K, Humbeeck J V, Aeenoudt E, Delannay F. Mater Charact, 1998; 40: 111
[17] Luo Z J, Wang L P, Wang M, Shen J C, Su H. Trans Mater Heat Treat, 2012; 33(2): 85 (Luo Zhijun, Wang Liping, Wang Meng, Shen Junchang, Su Hang. Transactions of Materials and Heat Treatment, 2012; 33(2): 85)
[18] Peng Y, Peng X N, Zhang X M, Tian Z L, Wang T. J Iron Steel Res Int, 2014; 21: 539
[19] Gadallah R, Fahmy R, Khalifa T, Sadek A. Int J Eng Technol Innov, 2012; 2(1): 1
[20] Yang C L, Lin S B. Fundamentals of Arc Welding. Harbin: Harbin Institute of Technology Press, 2003: 1 (Yang Chunli, Lin Sanbao. Fundamentals of Arc Welding. Harbin: Harbin Institute of Technology Press, 2003: 1)
[21] Zhang J C, Wang G R, Shi Y H, Zhong J G. Weld Pipe Tube, 2006; 29(3): 46 (Zhang Jianchun, Wang Guorong, Shi Yonghua, Zhong Jiguang. Welded Pipe and Tube, 2006; 29(3): 46)
[22] Steven W, Haynes A G. J Iron Steel Inst, 1956; 183: 349
[23] Okazaki Y, Ishida H, Suenaga K, Hidaka T. Weld J, 2012; 26: 593
[24] Huang W G, Xu R, Fang H S, Zheng Y K. J Iron Steel Res, 1997; 9(2): 3 (Huang Weigang, Xu Rong, Fang Hongsheng, Zheng Yankang. Journal of Iron and Steel Research, 1997; 9(2): 3)
[25] Song Y J, Lu J T, Liu J H, Shen L, Rao Q C, Zhou H J. Trans Mater Heat Treat, 1982; 3(1): 11 (Song Yujiu, Lu Jintang, Liu Jinghua, Shen Lian, Rao Qichang, Zhou Huijiu. Transactions of Materials and Heat Treatment, 1982; 3(1): 11)
[26] Yang Z G. Heat Treat Met, 2005; 30(1): 20 (Yang Zhigang. Heat Treatment of Metals, 2005; 30(1): 20)
[27] Madariaga I, Gutierrez I. Acta Mater, 1999; 47: 951
[28] Ma H, Chen J H, Zhu L. J Lanzhou Univ Technol, 1993; 19(2): 1 (Ma Hang, Chen Jianhong, Zhu Liang. Journal of Lanzhou University of Technology, 1993; 19(2): 1)
[29] Wu B Z, Xu Y J, An H L, Sun J T. Trans China Weld Inst, 2014; 35(4): 53 (Wu Bingzhi, Xu Yujun, An Hongliang, Sun Jingtao. Transactions of the China Welding Institution, 2014; 35(4): 53)

Note: Figure translations are in progress. See original paper for figures.

Source: ChinaXiv —Machine translation. Verify with original.