Thermomechanical Wear Testing of Metal Matrix Composite Cladding for Potential Application in Hot Rolling Mills

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Laser metal deposition (LMD) is utilized to clad the surface of a miniaturized test roll (Ø 40 mm) of tool steel. The cladding consists of two layers: a nickel alloy as intermediate layer deposited onto the surface of the steel substrate, and a metal matrix composite (MMC) as top layer consisting of spherical tungsten carbide particles embedded into the nickel alloy matrix. The thermomechanical wear behavior of the cladding is investigated on a test rig, where the test roll is pressed against an inductively heated load roll. Multiple test runs up to several hours simulating industrial loading conditions are performed. The presented testing procedure enables predicting the time-dependent abrasive wear behavior of the cladding, in particular for hot rolling mill applications. After testing for 8 h at temperature of 650 °C and at contact pressure of approximately 1 GPa, the maximum depth of the wear mark is about 0.12 mm. Partial cracking, debonding and dissolution of the tungsten carbide particles, as well as formation of iron and chromium oxides at the surface of the wear marks occur. However, as low abrasive wear is observed, the investigated MMC may potentially be applicable for cladding rolls in steel hot rolling mills.

1. Introduction

Hot rolling is the most important process for producing flat products in the modern steel industry. In particular, for hot rolling of steels, the rotating rolls must withstand severe wear caused by thermal and mechanical loadings. During the past decades, many new rolling technologies and roll materials were introduced to improve not only the cost and energy efficiency of rolling processes but also the quality of the products. Reducing wear to increase the lifespan of rolls is of main concern, as rolls and other mechanical mill equipment have always been a key factor influencing both the investment and the operating costs of steel mills.

Replacement and early breaking of rolls may represent up to 10% of the production costs. The loading mechanisms acting on the rolls and thus the requirements on the roll materials differ at various stages of the rolling line: at the initial stage strength and heat shock resistance are crucial, whereas the resistance against abrasive wear is essential at the following stages. The most serious wear mechanisms causing degradation of roll surfaces are abrasion (i.e., abrasive erosion), adhesion, thermomechanical fatigue, and tribochemical reactions (i.e., corrosion or oxidation).

Nowadays, rolls used in hot rolling mills consist of a comparatively soft ductile core, which is typically made of spheroidal or gray cast iron, and of a hard wear-resistant shell produced by spin casting. Three common types of shell materials are used: high-chromium (HiCr) steels, high-speed steels (HSS), and indefinite chill (IC) iron. Typically, rolls of HiCr and HSS are used in the early stands, whereas rolls of IC iron are used in the last two or three stands of hot strip finishing lines including six or seven stands. In particular, the growing use of HSS rolls and their continuous improvement since they were introduced in the late 1980s has gained increase in product quality and productivity, but decrease in production costs. In recent decades, numerous experimental studies have been performed to characterize the tribological behavior and the wear resistance of roll materials under thermomechanical loading. In particular, effects of different chemical compositions, thermal treatments, and process parameters during operation of the rolls (e.g., temperature, contact pressure) on both tribology and wear have been investigated. Discussing, the results of these studies would exceed the scope of this article; however, the ability of reducing abrasive wear of these “conventional” iron-based materials seems to be reaching its limit. Therefore, the present study aims to investigate an alternative material that could potentially be applied for hard-facing work rolls.
In recent years, the application of metal matrix composite (MMC) claddings has been increased in many industries for repairing and modifying surfaces exposed to harsh operating conditions. As these claddings have basically high resistance against abrasive wear, they are also expected to enhance the thermomechanical wear resistance of rolls used in hot rolling mills. Therefore, the basic idea was to deposit MMC layers, as already applied for tooling applications,[40–42] by laser cladding onto the surface of the rolls and to test the time-dependent wear behavior of the cladding layer by simulating industrial operating conditions at laboratory scale. Laser cladding, also known as laser metal deposition (LMD), is often used for depositing material layers onto a substrate. A laser beam is used for fusing powders preplaced on the substrate’s surface or continuously blown into the laser spot. Superficial melting of the substrate ensures strong metallurgical bonding of the cladding. Furthermore, the LMD process offers significant advantages regarding process flexibility and efficiency, limitation of base material distortions, and surface quality of the deposited material. Therefore, LMD has been widely applied for dimensional repair, surface modification, or additive generation of 3D structures.[43–45]

MMC claddings typically consist of crushed or spherical tungsten carbide particles embedded into an iron-, cobalt-, or nickel-based matrix. The microstructure and the wear behavior of these claddings are generally well studied at room temperature,[46–61] but only a few investigations focus on their particular wear behavior at elevated temperatures.[62–68] At room temperature, abrasive wear tends to decrease with increasing hardness of the MMC cladding, i.e., with increasing content or size of the hard particles inside the matrix, respectively. However, this overall trend does not necessarily confirm at elevated temperatures, where not only retained hardness but also the thermal stability of the microstructure of the matrix is of importance. Nevertheless, to the authors’ knowledge, no study on the wear behavior of MMC claddings considering both high temperatures and rolling-sliding conditions under high contact pressure has been conducted yet.

2. Materials and Methods

2.1. Producing the Cladding

In this work, a Trumpf DMD 505 LMD system equipped with a pulsed CO2 laser and a coaxial nozzle for powder feeding was utilized to clad the surface of a test roll (Ø 40 mm × 44 mm, EN 1.2343/X37CrMoV5-1). The cladding had two layers[40–42]: 1) a ductile nickel alloy interlayer deposited directly onto the steel substrate and 2) a nickel alloy matrix with embedded spherical tungsten carbide particles as hard top layer. The top layer was produced by adding powder of pure tungsten carbides (WC + W2C) while simultaneously remelting the nickel alloy, which was accomplished with the conventional laser cladding setup: the carbide powder was blown through a coaxial nozzle into the melt pool generated by the centered laser beam. To produce the MMC top layer without porosity and with homogeneous distribution of the carbide particles, the actual powder feeding rate must be carefully adapted to the volume of the melt pool, which is mainly influenced by the laser power, the spot size, the cladding speed, and the cladding track distance. The carbide content of the MMC top layer was about 50 wt%. Higher contents may promote both defect formation during production (cracks, pores) and brittle behavior during operation of the cladding.[41,42]

Table 1 shows the nominal compositions (wt%) of the powders used for laser cladding, and Table 2 summarizes the final process parameters that had been optimized for obtaining pore- and crack-free cladding layers.

Final machining of the hard MMC top layer for obtaining both the final roll diameter and the smooth roll surface was quite challenging. Wire cutting was tested and identified as option which is—in principle—applicable for machining the MMC top layer. However, restrictions regarding complex-shaped geometries and large dimensions and, particularly, the limited work speed will disqualify this process for large components. Hence, surface machining of the test roll was accomplished by means of conventional milling using cubic boron nitride (CBN) cutting inserts. Although a smooth surface of the roll was achieved, high wear of the milling cutters had to be accepted. Generating the top layer without defects was identified as critical process step influencing both machinability and mechanical properties of the cladding, because only a narrow process window exists for achieving an pore- and crack-free microstructure.

2.2. Testing the Cladding

2.2.1. Test Rig Design

Due to the high complexity of industrial hot rolling processes simulating equivalent conditions at laboratory scale is only feasible to a limited extend. Thus, configuration and conditions of the tests must be adapted in a way that approximates industrial
conditions as good as possible. Basic requirements for an appropriate hot rolling test rig were already defined a quarter of century ago.\textsuperscript{[3]} Several methods for experimental testing of tribology and wear behavior of potential roll and cladding materials have been used by researchers. The majority of these methods can be roughly classified as 1) pin-on-disc\textsuperscript{[54–26,33,36,37,49,51,63]} or ball-on-disc tests,\textsuperscript{[49,56–58,63]} 2) ring-on-disc tests with different setups\textsuperscript{[3,55,67]} 3) block-on-disc tests with\textsuperscript{[48,50,54,55,60,61,66,68]} or without\textsuperscript{[46,47]} addition of abrasive particles, and 4) disc-on-disc or roll-on-roll tests with two\textsuperscript{[3,4,13–19,23,27,28,34]} or even three\textsuperscript{[11]} discs or rolls, respectively. Among those test methods, in particular, the roll-on-roll tests are suitable to simulate the combined effects of relative motion (slip effects) and high contact pressure between the rolls of the rolling mill and the metal strip. Controlled heating of one of these rolls also allows simulating the effect of high strip temperatures in hot rolling.

The roll-on-roll test rig used in the present study was especially designed for investigating thermomechanical roll wear. The rig includes two rolls that are in contact while rotating about their individual axes: the test roll representing the work roll of a rolling mill and the heated load roll representing the hot metal strip. Main components are the central control unit, two independent drive units and the induction heating system. Figure 1 shows the CAD model of the test rig placed on the frame of a hydraulic press. During the wear test runs, the test rig is braced between the punch and the frame of this press. The IEW TH5 induction heating system includes a high-frequency generator, a heating station and an inductor. Each of the two drive units (upper drive unit and lower drive unit) consists of a three-phase synchronous motor with a flanged planetary gear box and a drive shaft with bearings mounted to a rigid frame of massive steel plates. As shown in Figure 1, one of the rolls is fixed to the end of each drive shaft: the test roll to the upper and the inductively heated load roll (Ø 80 mm × 10 mm, EN 1.4841/X15CrNiSi25-20) to the lower drive shaft. Tilting the entire upper drive unit presses the rolls together or detaches the rolls from each other, respectively. During the test runs, the surface temperature of the hot load roll is continuously monitored using a Sensotherm M323 pyrometer with an overall measurement range of 80–1200 °C.

Based on the temperature monitoring, inductive heating is controlled to keep the surface temperature of the load roll constant. The test rig was designed for maximum testing temperatures of 900–1000 °C, contact pressures of about 1.5 GPa, and slip of 20% between the rolls.

2.2.2. Testing Procedure

In the present study, the load roll had constant surface temperature of 650 °C after an initial heating-up period. The surface of the uncooled test roll also reached this temperature, because the surfaces of both rolls were in permanent contact. Work rolls in finishing stands of hot rolling mills typically attain temperature peaks of about 650 °C\textsuperscript{[5,38,39]} although flash temperatures may even be higher. To perform test runs at contact pressures equivalent to industrial rolling mills, the actual contact force applied between the rolls of the test rig, \( F \), was calculated according to the Hertzian contact theory.\textsuperscript{[50]} For linear contact of two parallel cylinders with different radii and different elastic properties, one can write\textsuperscript{[51]}

\[
F = \pi p_{\text{max}}^2 \frac{r_1 r_2}{r_1 + r_2} \left( \frac{1 - \nu_1^2}{E_1} + \frac{1 - \nu_2^2}{E_2} \right)
\]

(1)

where \( p_{\text{max}} \) is the contact pressure maximum occurring in the industrial hot rolling process simulated at the test rig. \( l \) is the length of the contact line between the two rolls, \( r_1 \) and \( r_2 \) are the radii of the rolls, \( E_1 \) and \( E_2 \) are the Young’s moduli, and \( \nu_1 \) and \( \nu_2 \) are the Poisson numbers of the roll materials. If both rolls consist of similar materials (\( E_1 \approx E_2 = E \) and \( \nu_1 \approx \nu_2 = \nu \)), e.g., when steel rolls for hot rolling of steel sheets are tested, and if the ratio of the main roll dimensions is \( r_1/r_2 = 4:2:1 \) (radius of load roll \( r_1 = r \), radius of test roll \( r_2 = r/2 \), length of contact line \( l = r/4 \)), Equation (1) simplifies to

\[
F \approx \frac{\pi}{6} p_{\text{max}}^2 \frac{(1 - \nu^2)}{E}
\]

(2)

Equation (2) was used with \( r = 40 \text{ mm} \), \( E = 210 \text{ GPa} \), and \( \nu = 0.3 \) for calculating the nominal contact force \( F = 3.6 \text{ kN} \).

![Figure 1. CAD model of the roll-to-roll test rig with main components.](image-url)
which was applied to obtain the nominal contact pressure maximum \( P_{\text{max}} = 1 \text{ GPa} \) between the rolls. Using Equation (2) was feasible because both the Young’s moduli and the Poisson numbers of steel and nickel alloy are quite similar. As the tungsten carbide particles are fully embedded into the nickel matrix, the matrix was assumed to dominate the elastic deformation behavior. The properties of the test roll (diameter, elastic properties) and the process parameters listed in Table 3 had been entered via intuitive visual touch screen panel directly at the control unit of the test rig.

Six test runs (TR0–TR5) were performed using always the same sample. Thus, the cladding layer was tested for 8 h in total, which corresponds to a rolling distance of about 7.5 km. During the test runs, actual values of the process parameters were monitored. Figure 2 shows the time-dependent contact pressure maximum calculated from the contact force monitored during the test runs. For each of the test runs the predefined range of 0.9–1.1 GPa was reached after an initial ramp-up period of 0.2 h. However, adapting manually the contact force was required for keeping the contact pressure within this range. Possible reasons for deviations are thermal expansion (increases the contact pressure) or progressive abrasive wear of the rolls (decreases the contact pressure), respectively.

Figure 3 shows the partially clad test roll exemplarily after test run TR0. The annealing colors at the clamping zone next to the clad zone indicate that the steel substrate was heated during testing, even though the hot load roll did not contact the substrate directly. Obviously, the wear zone can be clearly localized due to its different surface texture. However, macro- or microscopic changes of the cladding surface outside the wear test zone were not observed, neither after test run TR0 nor after the subsequent test runs TR1–TR5.

### 2.3. Investigating the Microstructure and the Wear Behavior

Basic characterization of the properties of the cladding layers included microstructure analysis, measurement of the layer thickness, and determination of the macrohardness using an automated EMCO-TEST DuraScan 70 G5 hardness tester. To investigate both the wear mechanisms acting on the roll during the test runs and the wear behavior of the roll material, different methods for microstructure analysis were applied to the worn surface. A TESCAN VEGA3 scanning electron microscope (SEM) equipped with an OXFORD X-Max 50 silicon drift detector (SDD) was utilized for analyzing chemical compositions of oxides and intermetallics. Backscattered electron (BSE) images and energy-dispersive X-ray (EDX) element maps were captured at three selected positions, P1, P2 and P3, as shown in Figure 3. For 3D surface imaging and for capturing the cross-section profiles of the wear mark at two positions, a KEYENCE VHX-6000 digital microscope was used. Based on these measurements, the local abrasion depth at the center of the wear mark was estimated after each of the test runs. Furthermore, the actual 3D geometry of the entire wear mark was captured using the optical measurement system GOM ATOS III Triple Scan after the last test run.

### 3. Results

#### 3.1. Hardness

Figure 4 shows exemplarily a typical hardness profile at the cross section of the cladding. Obviously, significant hardness differences between substrate, interlayer and top layer exist. The hardness of the top layer, i.e., the MMC layer consisting of the nickel alloy matrix with embedded tungsten carbide particles, is about 700 HV5 (≈60 HRC). That is actually higher than the hardness of the steel substrate, which is about 600 HV5 (≈55 HRC). In comparison, the hardness of the ductile interlayer, i.e., the carbide-free nickel alloy layer between the steel substrate and the top layer, is just about 350 HV5 (≈35 HRC). Note that the
comparatively thin top layer does not allow increasing the distance of the hardness imprints, because a sufficient number of imprints along the cross section of the cladding is mandatory for determining spatial hardness differences. Reducing the imprint size to increase the imprint distance (e.g., by reducing the load from HV5 to HV1) is also not expedient, as a certain imprint size is required for obtaining the representative hardness of the MMC, which is the “average” hardness of hard particles and soft matrix.

3.2. Surface Microstructure

3.2.1. Initial Microstructure

Figure 5 shows BSE images and EDX maps of the elements W, Ni, Fe, Cr and O captured at positions P1 and P2 inside the wear test zone and at position P3 outside the wear test zone. Images and maps were captured after the first test run TR0, because the surface conditions did not alter significantly during the subsequent test runs. It is obvious that the microstructure at position P3 was not influenced by the wear tests. The EDX maps show that W occurred exclusively inside the particles, Figure 5 (P3-W), whereas both Ni and Fe occurred exclusively inside the matrix, Figure 5 (P3-Ni) and (P3-Fe). However, thin boundary layers containing Ni and traces of Fe were observed around the particles. These interface layers consist of secondary intermetallic phases which form when carbides dissolve at high temperatures during the cladding process.\(^{[41,42,52,54]}\)

At first sight, Cr and O seem to be homogeneously distributed, Figure 5 (P3-Cr) and (P3-O), but closer examination reveals the content of O to be actually higher inside the particles than inside the matrix. Nevertheless, the comparatively low intensities of Figure 5 (P3-Cr) and (P3-O) indicate just low contents of Cr and O in the original cladding layer. The results of this EDX analysis show generally good qualitative agreement with the nominal chemical compositions, as shown in Table 1; however, according to the EDX maps, the actual Fe content of the matrix was much higher than expected.

3.2.2. Worn Microstructure

The microstructure of the cladding layer changed significantly inside the wear test zone at positions P1 and P2. Figure 5 (P1-BSE) and (P2-BSE) shows partial cracking, debonding, and dissolution of carbide particles. The EDX maps of W, Ni, Fe, Cr and O, captured at positions P1 and P2, also show significant differences compared with the maps captured at position P3. Both Fe and Cr, Figure 5 (P1-Fe), (P2-Fe), (P1-Cr) and (P2-Cr), accumulated at identical positions as O, Figure 5 (P1-O) and (P2-O). That strongly indicates the presence of Fe and Cr oxides at the surface of the test roll. These oxides were transferred from the surface of the load roll to the surface of the test roll. Formation and transfer of metal oxides are assumed to enhance at high temperatures or contact pressures, respectively.\(^{[23]}\)

Some of these dark gray oxides are exemplarily marked with red arrows in Figure 5 (P1-BSE) and (P2-BSE). Note that the formation of dense and compact oxide scales on roll surfaces would be basically a desirable effect in industrial hot rolling. In particular, protecting oxide scales forming at the surface of HSS rolls influence the friction conditions, reduce abrasive wear and thermal fatigue, and may improve the surface quality of the rolled products.\(^{[20,21,26,29,62]}\)

The current configuration of the test rig may foster the deposition of oxides on the surface of the test roll which has permanent contact with the load roll. Therefore, oxides forming at the surface of the load roll can be transferred to the surface of the test roll. In contrast, work rolls used in industrial hot rolling processes are in contact with changing counterfaces, as the metal strip passes the rolls continuously.

3.2.3. Particle Degradation

To illustrate the degradation process of the tungsten carbides, Figure 6 compares typical conditions a) after laser cladding, b) after machining, and c) after wear testing. Figure 6a shows details of a polished sample, whereas Figure 6b,c shows details of the surface of the test roll. It is obvious that the particles in the initial condition, Figure 6a, were free from internal cracks. The micrograph illustrates also the fragmented layer around the spherical carbide particles consisting of secondary intermetallic phases formed at high temperatures during the laser cladding process. The cluster of intermetallics in the upper right corner of the micrograph indicates that small carbide particles may even completely dissolve. While cracking of carbide particles was not observed directly after cladding, distinct signs of crack formation were determined after machining, as shown in Figure 6b. This indicates that machining processes (e.g., milling) induce particle cracking. Cracks of only few microns mainly originated from the intermetallic layer at the surface of the particles. Some small particles were even completely fractured. Furthermore, local plastic deformation during machining causes a sort of “smearing” of particle fragments and intermetallics, particularly in milling direction. Therefore, the particles lost their originally spherical shape. Closer examination of Figure 6b shows horizontal traces caused by the milling cutters; however, these traces disappeared completely during the wear tests, Figure 6c. During testing, the combined effects of rolling, sliding and high contact pressure at elevated temperature caused serious fracturing of the carbide particles, as exemplarily shown in Figure 6c. Fracturing facilitates debonding of the particles, as the fragments can detach more easily from the matrix.
Figure 5. BSE images and qualitative EDX maps of W, Ni, Fe, Cr and O for positions P1 (left column) and P2 (central column) inside the wear zone and P3 (right column) outside the wear zone.
3.3. Wear Characterization

After each of the six test runs, the surface topography and the cross-section profile of the wear mark were captured at two positions using the KEYENCE VHX-6000 digital microscope. Based on the time-dependent change of these cross-section profiles, the wear behavior was characterized. For that purpose, the measured profiles, Figure 7a, were initially smoothed using a median filter with a filter window of 2000 values. This operation removed both positive and negative peaks from the measured profile, Figure 7b, as these peaks represent local holes or oxide adhesions rather than circumferential grooves or elevations, respectively. The depth at the center of the wear mark, $\Delta r_2$, was then calculated as mean value of $i = 1...n$ depth increments, $\Delta r_{2,i}$, within the range of $\pm 1$ mm from the center of the wear mark:

$$\Delta r_2 = \frac{1}{n} \sum_{i=1}^{n(\pm1)} \Delta r_{2,i} \quad (3)$$

In addition, the abrasive volume loss $\Delta V$ of the cladding layer was estimated after each test run. Therefore, the circumference of the test roll was multiplied with the cross-section area of the wear mark. This area is enclosed by the smoothed cross-section profile and by the original contour of the roll. It was calculated as sum of $j = 1...N$ area increments within the total width of the wear mark of $\pm 6$ mm. In Equation (4), $r_2 = 20$ mm is the radius of the test roll, $\Delta x = 0.71 \mu$m is the width increment, and $N \approx 17000$ is the total number of area increments along the width of the wear mark.

$$\Delta V = 2\pi r_2 \sum_{j=1(-6)}^{N(\pm6)} |\Delta r_{2,j}\Delta x| \quad (4)$$

Figure 6. Carbide particles a) after cladding, b) after machining, and c) after the first test run.

Figure 7. a) Wear profiles measured, b) wear profiles smoothed, c) depth at the center of the wear mark, and d) loss of cladding material.
The results of the calculations are illustrated in Figure 7c,d. Each point shown in the diagrams was calculated based on a cross-section profile measured after test runs TR0–TR5, as shown in Figure 7a. Both the depth at the center of the wear mark (center depth) and the loss of cladding material due to abrasion (volume loss) can be described almost identical using simple power-law functions. Based on these functions, both wear depth and volume loss beyond the duration of the tests can be predicted. For example, operating the cladding for 10 h under the given testing conditions would result in a volume loss of about 160 mm³ due to abrasive wear. The center depth of the wear mark would be about 0.15 mm, which is approximately 1/3 of the total thickness of the MMC top layer, as shown in Figure 4. After the initial nonlinear period, the cladding had an almost constant wear rate of 18 mm³ h⁻¹.

Figure 8 shows the topography of the wear mark after the test runs TR1, TR3 and TR5, i.e., after 2, 4 and 8 h of testing. Note that each of these three topography maps was captured at the identical position of the roll. Thus, the maps illustrate the development of the wear mark over the testing time. Figure 7 and 8 show that after 8 h of testing, the maximum depth at the center of the wear mark was about 0.12 mm. Abrasion of the MMC top layer was relatively homogeneous during the early test runs TR1 and TR3. In Figure 8a,b, but transverse grooves across the wear mark were observed after the last test run TR5, as shown in Figure 8c. These grooves indicate enhanced local abrasion. As they were oriented almost parallel to the central axis of the roll, they are assumed to form predominantly at the overlap of the cladding tracks where lack of hard-facing particles could occur. The cladding tracks as well as the final milling tracks were oriented parallel to axis of the roll, i.e., the cladding tracks were oriented perpendicular to the rolling direction.

Figure 9 shows the 3D representation of the complete test roll geometry captured after test run TR5 using the GOM ATOS III Triple Scan system. The measured geometry data were superimposed with an idealized cylinder (Ø 40 mm × 44 mm) to quantify the depth of the wear mark. Obviously, the maximum depth determined with the optical measurement system corresponds well to the maximum depth determined with the digital microscope. Closer examination of Figure 9 also reveals diffuse bands running across the wear mark and representing grooves of enhanced local wear.

4. Conclusions

The present work investigates the thermomechanical wear behavior of a MMC cladding using a roll-on-roll test rig designed to simulate conditions occurring in industrial hot rolling mills. The cladding consists of two layers: a comparatively soft nickel alloy as ductile interlayer and a nickel alloy matrix containing tungsten carbide particles as hard top layer intended to improve significantly the wear behavior of the rolls. Several test runs with 8 h of total test duration were performed with contact pressure of 1 ± 0.1 GPa, 10% slip, and 650 °C surface temperature. Based on the obtained results the following conclusions can be drawn: 1) A dense MMC top layer without pores or cracks was produced with the laser cladding process. However, only a narrow process window exists for obtaining defect-free cladding layers. This window must be identified depending on the actual material combination (substrate/matrix/particles). Final machining of the cladding surface caused high wear of the milling cutters. Generating an almost defect-free MMC top layer was identified as crucial for achieving good machinability and excellent mechanical properties of the cladding. 2) The designed test rig was successfully utilized for simulating the rolling-sliding contact as it occurs in rolling mills between the work rolls and the hot metal strip. However, to improve the comparability between testing conditions and industrial conditions, optimizations and modifications of the test rig design (e.g., third roll to consider the influence of support rolls, droplet lubrication, cooling with compressed air, compensation of thermal expansion) are in progress. 3) Microstructure investigations after testing revealed partial cracking, debonding, and dissolution of carbide particles. These phenomena were mainly induced by the combination of high contact pressure and slip between the rolls. Although these kinds of particle deterioration were already observed after the first test run, they did not cause a rapid decay of the wear resistance of the MMC cladding in the subsequent test runs. 4) The presence of both Fe and Cr oxides at the surface of the test roll was observed. These oxides formed at the load roll and were transferred to the test roll where they locally adhered. As no closed oxide scale formed, the oxides did not significantly
contribute to the tribology or to the wear behavior of the MMC cladding. 5) Under the given testing conditions, the cladding showed high abrasive wear resistance. Operating the cladding for 10 h would result in a maximum depth of the wear mark of about 0.15 mm, which is approximately 1/3 of the total thickness of the MMC top layer. Both the depth at the center of the wear mark and the loss of cladding material due to abrasion can be described using simple power-law functions. After the initial nonlinear period, the MMC cladding showed an almost constant wear rate of 18 mm$^3$ h$^{-1}$. 6) Based on the results of this study, the MMC cladding is recommended for further testing in pilot rolling mills. The abrasive wear resistance of the cladding under thermomechanical loading, i.e., under rolling and sliding at both high contact pressure and temperature, seems to be excellent. However, the surface quality of the rolled product is still an open question, and final machining will be challenging due to the hardness of the cladding.

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Conflict of Interest
The authors declare no conflict of interest.

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