Effect of cold deformation on the stress corrosion cracking resistance of a high-strength stainless steel

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ABSTRACT

The resistance to chloride-induced stress corrosion cracking was investigated on a high-strength CrNiMnMoN austenitic stainless steel in the hot-rolled and in different cold-drawn states. The resistance against chloride-induced stress corrosion cracking was determined by slow strain rate tests in different chloride containing solutions at elevated temperatures. A fracture analysis was carried out using scanning electron microscopy. Improved resistance is obtained by the formation of deformation-induced twins. In addition, synchrotron X-ray diffraction measurements show full austenite stability during all cold-drawing steps.
Introduction

Austenitic stainless steels are widely used in the offshore oil and gas industry. The most commonly used steels are from the CrNiMo stainless steel class. These are characterized by excellent corrosion resistance, but their low strength often limits their use. As an alternative, CrMnN stainless steels are frequently used, which are characterized by higher strength but significantly lower corrosion resistance. In particular, their resistance to chloride-induced stress corrosion cracking (Cl-SCC) is clearly unsatisfactory. In order to be able to guarantee the highest corrosion resistance as well as high strength levels, a new material class of CrNiMnMoN stainless steels is preferred nowadays [1, 2]. The excellent resistance to hydrogen- and chloride-induced stress corrosion cracking of a hot-rolled CrNiMnMoN wire was shown in a previous paper [3]. In this study, special attention is paid to the effect of a further increase in strength on the resistance to Cl-SCC. The central point is to clarify whether and from which degrees of deformation upwards an increased tendency to embrittlement of the material occurs.

The most pronounced feature of cold forming is the increase in dislocation density. Other effects that can occur during cold forming are grain refinement, the formation of voids and, in the case of austenitic steels, deformation-induced twinning and deformation-induced martensite formation are also possible. These effects lead to an increase in strength, but except for grain refinement, also to a decrease in ductility. In particular, deformation-induced martensite formation can severely reduce the elongation to fracture of an austenitic steel [4, 5]. The stacking fault energy determines whether deformation-induced martensite forms during cold deformation or deformation-induced twinning occurs. Deformation-induced martensite forms at stacking fault energies below 20 mJ m$^{-2}$. Deformation-induced twinning is more likely to occur at stacking fault energies between 20 and 45 mJ m$^{-2}$ [6, 7]. At stacking fault energies above 45 mJ m$^{-2}$, dislocations hardly split at all and dislocation slip occurs instead of twinning [7, 8]. Heavy cold deformation also leads to the introduction of high residual stresses, whereby tensile residual stresses are usually critical for material failure [9].

Resistance against pitting is strongly related to resistance against SCC, as pits can act as crack initiators [10, 11]. Nitrogen is known to improve repassivation behavior in pitting corrosion by formation of ammonia. On the same side repassivation kinetics is of highest importance in the slip-step dissolution model to retard dissolution during SCC formation. Consequently, N can be expected to
improve SCC resistance [12]. The same positive effects appear for Cr and Mo in both types of corrosion (pitting and SCC). Nickel is known to improve SCC resistance due to its noble electrochemical potential and its uniform corrosion rate in chloride containing solutions. In addition, Ni increases the stacking fault energy and promotes fine slipping [13–17]. However, it decreases according to Speidel the PREN. Speidel has found a negative factor of 0.25 in the PREN for every percent of Ni added to a stainless steel [18].

Several effects are claimed to be responsible for the change of Cl$^-$ SCC resistance of cold deformed stainless steels. Fine grains tend to improve the resistance to Cl$^-$ SCC due to the increase in toughness [16, 19]. The morphology of the grain boundaries shows that cracks are more likely to occur along high-angle grain boundaries and less likely along low-angle grain boundaries or special high-angle grain boundaries such as a twin boundary. A general dependence is difficult to determine due to the strong influence of the preferred crack propagation (transgranular or intergranular) [12, 20, 21].

The presence of deformation-induced martensite significantly lowers the resistance to Cl$^-$ SCC as reported in Refs. [5, 22]. Deformation-induced martensite has a lower pitting potential and can act as SCC initiation sites for different material—environment combinations [23, 24]. Deformation-induced twinning, however, should be favorable in comparison with deformation-induced martensite formation, because of a better resistance against SCC, but has been reported for very few material-environment combinations up to now [20, 21].

The introduction of tensile residual stresses during cold working leads to reduced resistance against SCC, resulting in the appearance of either low residual stresses (5% of the yield strength) or high residual stresses (up to 30% of the yield strength) [9, 15]. Existing tensile residual stresses significantly reduce the ability of a material to withstand external loads and increase the probability of failure [25].

Another fact is that in conventional stainless steels damage to the passive layer and preferential pitting corrosion can occur even at low degrees of deformation due to the increased defect density. High nitrogen steels (0.9 wt% N) suffer damage of the passive layer only above 60% of cold deformation [26–28].

In summary, published data show that resistance to SCC is significantly reduced at low degrees of deformation, whereas higher degrees of deformation, i.e., above 30%, usually lead to improved resistance. The minimum is usually in the range between 5 and 30% of cold deformation, but it can be reduced more drastically, if sufficient high degrees of deformation are applied [8, 17, 20, 29]. It was shown by Cigada et al. [5], that there has to be made a distinction between crack initiation and crack growth. At low deformation levels, crack initiation is accelerated by the appearance of defects that can act as crack initiators, and in areas of low cold deformation, accelerated crack initiation dominates [5, 30]. The formation of dislocation structures, subgrains or transformation-induced twinning occurs to a greater extent at higher degrees of deformation [30, 31]. An increase in defect density in this range has only a minor effect on crack initiation, but influences crack growth much more. Crack growth is delayed by these obstacles [20, 32, 33]. Particularly special grain boundaries such as twins or other coincidence site lattice boundaries have an increased corrosion resistance and can therefore improve the SCC resistance [34]. At extremely high degrees of deformation, however, there is often a limit on this effect, as other negative effects, such as increased residual stresses or the appearance of strong textures that facilitate crack growth, then come to the foreground [15].

So far, no studies on the SCC resistance of cold-drawn CrNiMnMoN stainless steel wires have been carried out. The present work investigates the resistance of a high nitrogen alloyed stainless steel wire in different degrees of cold deformation with special focus on microstructural changes during the drawing process and its influence on the SCC resistance. In addition, individual conditions of cold working and the prevailing effects on Cl$^-$ SCC resistance are discussed in detail.

**Experimental**

**Materials**

Starting with a hot-rolled X3CrNiMnMoN 27-14-6-3 stainless steel with 0.7 wt% N, different cold-drawn conditions of this stainless steel wire were examined. Cold-drawing was carried out with conventional drawing dies in five drawing steps. The initial wire
diameter was 9 mm and will further be referred to as hot-rolled. The production of hot-rolled wire is described in Ref. [3]. The mechanical properties of all tested material conditions are summarized in Table 1.

**Materials characterization**

The multi-scale characterization of samples in different drawing stages includes optical microscopy, cross-sectional synchrotron x-ray diffraction (CSmicroXRD) and electron backscatter diffraction (EBSD).

The determination of the grain size was done with an Olympus AX70 optical microscope. Steel wire specimens were cut, embedded, polished and cathodically etched in a 10 wt% oxalic acid. The determination of the average grain size was conducted on the etched cross sections according to ASTM E112.

CSmicroXRD experiments were performed to determine depth gradients of phases and residual x-ray elastic strains. The samples were measured at the side-hutch P07B of the high-energy materials science beamline (HEMS) of DESY at the storage ring PETRAIII in Hamburg, Germany [35]. A beam energy and a beam size of 87.1 keV and $500 \times 100 \mu m^2$, respectively, enabled to use a transmission diffraction geometry. The spatial resolution was determined by the choice of a scanning increment of 100 $\mu m$ in height. The diffracted photons were collected by a two-dimensional digital X-ray flat panel PerkinElmer detector, type-XRD1621, with a pixel pitch of 200 $\mu m$. A LaB$_6$ standard was used to calibrate the distance between the sample and the detector. In a next step, collected data were processed by using the Python software package pyFAI [36]. Thus, each 2D-pattern was radially integrated in azimuthal sections, each of $\Delta \delta = 10^\circ$ in width, in order to quantify the direction dependent lattice spacing for 36 diffraction vector orientations. Then, the austenite 311-hkl peak was fitted using a Pseudo–Voigt function and residual stresses were evaluated from measured x-ray elastic strains. X-ray elastic constants of $S_1 = -1.3780 \times 10^{-6} MPa^{-1}$ and $S_1 = 6.565 \times 10^{-6} MPa^{-1}$ [37] were calculated using single-crystal elastic constants determined by the Eshelby-Kroener grain-interaction model [38]. A detailed description of the evaluation is provided in Refs. [39–41].

For EBSD analysis a FEI Versa 3D Dual Beam focused ion beam work station with attached scanning electron microscope (SEM), equipped with a Hikari XP EBSD camera, was used. The measurement was conducted on cross sections of the X3CrNiMnMoN 27-14-6-3 stainless steel using a step size of 200 nm for the hot-rolled condition and 100 nm for the cold-drawn states. The data evaluation was done with the Advanced OIM software package, version 7.3.0. For visualization a grain dilation clean-up with a grain tolerance angle of 2.5$^\circ$ and a minimum grain size of 1 $\mu m$ was applied. The metallographic sections were subjected to a final polishing step with non-drying fumed silica suspension OP-S for 8 min for EBSD analysis.

**Corrosion testing**

Slow strain rate tests (SSRT) were performed on non-standard tensile specimens with a gauge diameter of 3 mm and an initial gauge length of 25.4 mm with a cross-head speed of 0.003 mm min$^{-1}$ ($\sim 2 \times 10^{-6}$ s$^{-1}$) for the hot-rolled and 0.0003 mm min$^{-1}$ ($\sim 2 \times 10^{-7}$ s$^{-1}$) for the cold worked conditions. For the SSRTs, in principle one test was carried out per

| Table 1 | Degree of cold reduction (CR), yield strength (YS), ultimate tensile strength (UTS), elongation to fracture (A) and the wire diameter (d) of the investigated X3CrNiMnMoN 27-14-6-3 stainless steel |
|---------|---------------------------------------------------------------------------------------------|
| Material condition | Degree of cold reduction (CR) (%) | Yield strength (YS) (MPa) | Ultimate tensile strength (UTS) (MPa) | Elongation to fracture (A) (%) | Wire diameter (d) (mm) |
| Hot-rolled | 0.0 | 620 | 1050 | 49.19 | 9.0 |
| Cold-drawn | 20.0 | 1135 | 1303 | 18.73 | 8.1 |
| Cold-drawn | 36.0 | 1383 | 1637 | 8.77 | 7.2 |
| Cold-drawn | 48.7 | 1656 | 1729 | 6.88 | 6.4 |
| Cold-drawn | 59.1 | 1674 | 1881 | 6.54 | 5.8 |
| Cold-drawn | 67.3 | 1838 | 1922 | 2.56 | 5.2 |
material—environment combination. Selected conditions were tested three times to ensure repeatability. Detailed informations on the experimental setup are given in Ref. [3]. Three different aggressive media were chosen as testing conditions: (1) 5 wt% NaCl, buffered with NaHCO₃ and HCl to pH 3.5, at 80 °C, (2) the more aggressive 43 wt% CaCl₂ solution at 120 °C and, (3) the most aggressive solution a 42 wt% MgCl₂ solution at 120 °C. Measurements were also performed in glycerine at 80 °C, buffered with NaHCO₃ and HCl to pH 3.5, at 80 °C.

To determine the so-called resistance index for the elongation, which compares the elongation under inert conditions, we define the elongation ratio as:

\[
\text{Elongation Ratio (RE)} = \frac{E_a}{E_i}
\] (1)

Reduction of Area Ratio (RRA) = \frac{RA_a}{RA_i}

The resistance values RE and RRA are very useful for comparing the resistance of cold-drawn wires. Other evaluation methods such as the proportion of embrittled area or the number of secondary cracks are not suitable, as cold-drawing can cause a change in the fracture morphology.

Results

Materials properties

Cold-drawing results in significant changes regarding microstructure and mechanical properties. Figure 1 presents the grain size according to ASTM E112, the fraction of two types of grain boundaries as well as an inverse pole figure (IPF) maps of the samples.

The average grain diameter in the hot-rolled condition is 11.4 μm, while the cold-drawn conditions show average grain diameter in the range between 5.3 and 7.0 μm. Cold working leads to an increased defect density (dislocations, vacancies) and to the formation of substructures by LAGBs associated with a decrease in grain size. The reduction in grain size occurs through the rotation of individual grains during the drawing process and the resulting formation of smaller grains as the wire diameter decreases.

At higher degrees of deformation, significant changes could be observed with respect to an elongated shape of grains along the drawing axis. An increased defect density, e.g., geometrically necessary dislocation (GND), can be observed in a rising number of color transitions in and around the grains at the IPF maps. The percentage of low-angle grain boundaries (LAGBs) increases to 62% after the first cold-drawing step and remains relatively constant during further deformation.

Because the employed EBSD system was not able to fully detect the typically small deformation induced twins [42], a detailed analysis of the twins is restricted to a specific section from the scan of the material, namely with 67.3% applied degree of deformation. Figure 2a, b shows the respective IPFs after rotating the data map for 90°, i.e., perpendicular to the drawing direction. While the blue areas account for grains in the (111) orientation, the red areas are those grains responsible for the second present fiber (100). Also visible within the (111) oriented grains are sharp red lines, showing a (100) texture as well. These sharp red lines are of about a 58° misorientation (highlighted in Fig. 2c) to their belonging (111) grains and therefore can be described as austenitic twins [43, 44]. When comparing these results with the IPFs from Fig. 1, it can be stated that twinning occurs preferably in the (111) oriented grains when the CR is ≥ 48.7%.

In Fig. 3, the pole figures (PFs) of {100}, {101} and {111} lattice planes, derived from EBSD data, allow to observe the texture evolution for increasing degrees of cold deformation. The texture intensity is indicated in multiples of a random distribution (MRD) in the range between 0.000 and 6.000. The PFs of the recrystallized starting material, i.e., the hot-rolled wire condition (see Fig. 3a) shows no pronounced texture. Comparing Fig. 3a–d, it can be stated that a (111) fibre texture develops when the degree of cold deformation is ≥ 20%, increasing with even higher degrees of deformations of up to 67.3%. Additionally, a weaker (001) fibre texture is present in the cold deformed states, see Fig. 3b–d, when comparing the 001-PFs. The maximum MRD value for the (111) fibre reaches a maximum of 5.6, while the maximum of the (100) fibre lies at 1.7 at 67.3% degree of cold deformation.
Austenite stability in samples of each cold-drawing stage was examined using CSmicroXRD. Figure 4 shows a stack of diffractograms along the entire cross section of the specimen in the hot-rolled and most deformed state (67.3% CR). As no harmonic rejection optics was installed in the experimental setup, higher harmonics of the used wavelength, fulfilling Bragg’s law with a higher order reflection cause additional weak (austenite) peaks in the phase plot of the hot-rolled condition.

As no martensite can be detected by CSmicroXRD, see the phase plot of the 67.3% condition, which corresponds to the highest degree of deformation, proves the high stability of the austenite. Especially in the near-surface region where maximal tensile residual stresses of ~1121 MPa could be evaluated, the stability of the austenite is a crucial requirement for SCC resistance. The peak broadening up to depths of ~1 mm indicates a grain refinement as well as the presence of second- and third-order residual stresses. The distribution of residual stresses in Fig. 4 indicates that maximal axial residual stresses in the material are induced near the surface, then decreasing toward the centre of the specimen. In the hot-rolled material, there are only minor residual tensile stresses, whereas at higher degrees of cold forming, the level of residual stresses increase.

Figure 5 shows the comparison of maximum tensile stresses in the wire and the yield strength of the material. The results indicate a strong correlation of yield strength and maximal residual stresses for all degrees of cold deformation. The maximum residual stresses along the wire were evaluated from the results of the residual stress profiles. The measurements of the residual stresses along the wire diameter of the respective states are single measurements due to limited beamtime.

![Figure 1](image1.png)  
**Figure 1** Grain size of the different conditions of the X3CrNiMnMnN 27-14-6-3 stainless steel according to ASTM E112; fraction of low-angle grain boundaries (LAGB) with a grain misorientation angle $\varphi < 15^\circ$ and the fraction of high-angle grain boundaries (HAGB) with a grain misorientation angle $\varphi > 15^\circ$ as well as selected inverse pole figure maps of the longitudinal cross sections.

![Figure 2](image2.png)  
**Figure 2** IPFs perpendicular to the drawing axis of the 67.3% cold-drawn material.
Corrosion resistance

The resistance indices as a function of deformation for the different test conditions are shown in Fig. 6. It should be noted that the strain rate for the hot-rolled condition was chosen to be $2 \times 10^{-6}$ s$^{-1}$ due to the large elongation to fracture, while the cold-drawn conditions were tested with a strain rate of $2 \times 10^{-7}$ s$^{-1}$. The buffered 5 wt% NaCl solution at 80 °C shows no detectable embrittlement of the material, while the most aggressive testing environment, i.e., 42 wt% MgCl$_2$ at 120 °C, leads to significant embrittlement for all material conditions. If considering these two test solutions, no significant influence of the degree of cold forming on the course of the curves can be detected. Therefore there is the need to test the material with a medium aggressive solution to distinguish between the different material conditions. Testing in 43 wt% CaCl$_2$ shows that cold-drawing increases the tendency to Cl$^-$ SCC. At 20% cold forming, the resistance to Cl$^-$ SCC is significantly reduced, which increases slightly with further cold forming and only decreases to resistance values below those of the 20% state at degrees of cold deformation above 59.1%.

Figure 7 shows the stress–strain curves of the different states in 43 wt% CaCl$_2$ solution. The hot-rolled material has a high resistance against Cl$^-$ SCC, while the cold-drawn states show significant reduction in fracture elongation in the respective solution. The state with 48.7% CR loses the least fracture elongation among these states, while the states 59.1% CR and 67.3% CR hardly show any resistance in the testing solution.

SEM images of the fracture surfaces of SSRT specimens tested in CaCl$_2$ are shown in Fig. 8. The SSRT of the hot-rolled condition shows no embrittlement and a clearly ductile failure is recognisable by the appearance of pronounced dimples [3]. In the same test medium, at 20% cold deformation, there is clearly recognisable embrittlement of the material (Fig. 8a). At the fracture surface, a

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**Figure 3** Pole figures of {001}, {101} and {111} lattice planes of different materials conditions of the X3CrNiMnMoN 27-14-6-3 stainless steel, measured by EBSD.
A distinction can be made between SCC and the residual fracture. The brittle failure (Fig. 8b) is a mixed form of intergranular and transgranular cracking. Furthermore, fracture necking could be determined to be very low and almost no deformation lines are visible on the sample surface. Figure 8c indicates a larger fracture necking for 36% cold deformation and deformation lines are visible in this case (white arrows). The fracture surface in Fig. 8d shows almost pure transgranular SCC for the 36% cold deformation case. The area of residual fracture is similar to the previous sample. At 59.1% cold deformation, there is obvious embrittlement, which is shown by the low reduction of area and the large reduction of strength. 

**Figure 4** Residual stress profiles of hot-rolled, 20% CR and 67.3% CR conditions were assessed by cross-sectional synchrotron XRD; phase plots of the cold-drawn X3CrNiMnMoN 27-14-6-3 stainless steel in the hot-rolled and 67.3% CR conditions indicate the stability of the austenite.

**Figure 5** Residual stresses of the respective deformation state evaluated over the entire wire diameter in relation to the respective yield strength.

**Figure 6** Influence of the degree of cold deformation on the resistance to Cl− SCC of the X3CrNiMnMoN 27-14-6-3 stainless steel given by the resistance indices of the elongation RE and the reduction in area RRA. The strain rate was $2 \times 10^{-6} \text{ s}^{-1}$ in the hot-rolled condition (0% degree of cold deformation) and $2 \times 10^{-7} \text{ s}^{-1}$ for the different cold worked states.
residual fracture area (Fig. 8e). Obviously, the formation and propagation of small cracks are sufficient to cause the material to fail. It has to be stated that this is related to the lower fracture strain of the initial state as well. The fracture surface in Fig. 8f also shows increasing portions of intergranular failure. The fracture appearance shows greater portions of intergranular cracks for the 20% cold-drawn and the 59.1% cold-drawn condition than for the 36% cold-drawn condition.

Discussion

If the resistance of three different media is differentiated, it can be seen that the 5 wt% NaCl solution at 80 °C does not lead to any embrittlement at any degree of deformation. The repassivation of the material is so strong at these temperatures that no attack occurs. With the most aggressive medium, the 42 wt% MgCl₂ solution at 120 °C, it appears that there is a strong breakdown of the passive layer in this medium and thus only a minimal residual resistance is achieved in all deformation levels. Small differences for different degrees of cold deformation are attributed to scatter of results. 43 wt% CaCl₂ solution at 120 °C, however, is an electrolyte just aggressive enough to initiate SCC, but also mild enough to differentiate between different degrees of cold deformation. It is a general principle in corrosion science that mild media result in a better distinction of corrosion properties of different states of material (such as different degrees of cold deformation). The very high PREN value of this material and the SEM images shown do not suggest that breakdown of the passive layer occurs in 43 wt% CaCl₂ solution. This also fits well with previous results showing a correlation between pitting and SCC resistance in materials with lower PREN values [10, 11].

The overall trend shows a higher tendency of embrittlement for higher degrees of deformation. This observation can be explained by the presence of rather high residual stresses and a higher defect density at higher deformed states. Also Refs. [9, 15] found that a high plastic deformation is detrimental on Cl⁻-SCC resistance. The residual stresses rise continuously for higher degrees of deformation and exceed 1000 MPa of tensile residual stresses for the two highest degrees of deformation.

The texture analysis revealed typical deformation textures for austenitic wire when deformed under uniaxial stress, e.g., via cold-drawing. As reported by Engler et al. [45] for face-centered cubic materials with low stacking fault energy in general or by Hwang [46], specifically for austenitic steel wire, cold-drawing leads to the evolution of two fiber textures with the ⟨111⟩ and ⟨100⟩ direction parallel to the drawing direction. The stronger component, which turned out to be the ⟨111⟩ fibre, is documented to have an enhancing effect on the twinning rate [46]. Comparing the results from texture analysis it can be summarized that the strengthening of the ⟨111⟩ fibre
at increasing degrees of cold deformation goes in hand with the enhanced deformation-induced twinning, observed after applying CR of ≥ 48.7%. An exception to the trend of decreasing resistance with increasing cold working can be observed in the range between 20 and 48.7% CR. In the material X3CrNiMnMoN 27-14-6-3 the formation of deformation-induced twin boundaries at CR of ≥ 48.7% could be detected, while there is no occurrence of deformation-induced twinning in the 20% cold formed state as seen in Fig. 9. While for the hot-rolled and the 20% CR conditions, annealing twins are randomly distributed throughout the recrystallized microstructure, within the higher deformed states, i.e., 48.7% and 67.3% CR, a highly dense and parallel alignment of twins within the elongated grains was observed. A similar observation was made by Barbier et al. [42] for a twinning-induced plasticity steel. At low degrees of deformation, mechanical twinning could be detected by means of transmission electron microscopy.

Figure 8 Selected SEM fracture surface images of SSRT specimens of X3CrNiMnMoN 27-14-6-3 stainless steel tested in 43 wt% CaCl2 solution at 120 °C: a 20% cold deformation overview, b 20% detail, c 36% cold deformation overview, d 36% detail, e 59.1% cold deformation overview, f 59.1% detail.
microscopy, while increased twinning formation at higher degrees of deformation could be resolved by EBSD.

There is a strong increase in twin boundaries at higher levels of deformation, while there is no increase from 0 to 20% cold deformation, which leads to the assumption that the occurrence of deformation-induced twinning increases the resistance to Cl$^-$SCC. Any grain boundary is an obstacle for crack propagation and at grain boundaries dislocations pile-up. The results in Fig. 1 shows that grain size decreases only to a small extent by cold deformation, however the density of twin boundaries increases sharply between 20 and 48.7% CR. Consequently, there is an increase in dislocation pile-ups at these twin boundaries hindering crack propagation [47]. In addition, Mukai et al. [48] described that cracks, which can easily grow in the $<$110$>$ direction, are hindered by twin boundaries and micro-branching of the cracks occurs. This micro-branching subsequently leads to a delay in crack propagation and thus to improved SCC resistance. For a twinning-induced plasticity steel in a caustic solution with pH 12.4, it has been shown that twins are preferential sites for localized corrosion, but due to their good ability to deform, instead of suffering decohesion, they improve the resistance against SCC [49]. Figure 10 summarizes

**Figure 9** SEM images with indication of twin boundaries in the range of $\phi = 55 - 65^\circ$ of different conditions of the X3CrNiMnMoN 27-14-6-3 stainless steel, measured by EBSD.

**Figure 10** Influence of deformation-induced twinning on the resistance to Cl$^-$SCC for a cold-drawn X3CrNiMnMoN 27-14-6-3 stainless steel. The standard deviation is 0.05 and was determined by three parallel samples of the condition CR = 48.7% in 43 wt% CaCl$_2$, at 120 °C.
the influence of deformation-induced twinning on the resistance to Cl\textsuperscript{-} SCC.

In general, the SCC resistance decreases with higher degrees of deformation due to the greatly increased defect density and higher residual stresses. Muraleedharan et al. [30] have shown in their work for stainless steels type 304 and 316 a decrease in SCC resistance with increasing degrees of cold deformation up to 20% CR. They mention that cold deformation leads to the availability of a large number of defects which can act as crack-nucleation sites. This would result in a decrease in crack initiation time and would explain the strong decrease in the SCC resistance of the 20% CR material in 43 wt% CaCl\textsubscript{2} solution. The large increase in defect sites within the first cold-drawing step is shown in Fig. 1. In our work we see similar behavior and we conclude that there is no crack initiation in the hot-rolled material and therefore excellent resistance in CaCl\textsubscript{2}. In cold-drawn material, defects are introduced in such a large number that crack initiation is quite easy and can grow quite unhindered through the material. As the degree of deformation increases, simple crack initiation at the defects is still possible, but the cracks are increasingly blocked and branched by the twin boundaries. This results in increased resistance to Cl\textsuperscript{-} SCC up to 50% CR.

The above mentioned fibre texture at higher degrees of deformation enhances the formation of mechanical twins with increasing degrees of deformation. The mechanical twins lead to a significantly improved resistance to Cl\textsuperscript{-} SCC while its resistance decreases again at very high degrees of cold deformation. At very high degrees of deformation above 50%, there is an increasing propagation of the crack in the wire drawing direction.

**Conclusions**

The Cl\textsuperscript{-} SCC resistance of a cold-drawn stainless steel wire is improved through deformation-induced twinning in the range between 20 and 48.7% CR. In general the Cl\textsuperscript{-} SCC resistance decreases for higher degrees of cold deformation.

The following conclusions can be drawn for the use of cold-drawn high nitrogen stainless steel wires:

- Deformation-induced twinning improves the resistance against Cl\textsuperscript{-} SCC. Higher beneficial effect is obtained between 36 and 48% of cold deformation.
- High residual stresses and a higher defect density decrease the resistance to Cl\textsuperscript{-} SCC and very high residual stresses are predominant at degrees of cold deformation higher than 50%.
- The (111) fibre texture of the material enhances the formation of mechanical twins above 20% of cold deformation.
- Austenite stability is required to achieve a good resistance against Cl\textsuperscript{-} SCC for cold-drawn stainless steel wires.

The best resistance with the highest possible strength can therefore be achieved by the X3CrNiMnMoN 27-14-6-3 at degrees of cold deformation in the range of 36–48%.

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**Authors contribution**

MT Conceptualization, Methodology, Writing—Original Draft. AJ Investigation, Visualization. SCB Investigation, Visualization. AK Conceptualization,
Resources. GM Conceptualization, Writing—Review and Editing, Supervision.

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Data availability statement

The raw/processed data required to reproduce these findings cannot be shared at this time as the data also forms part of an ongoing study.

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