# A 01 ***

*Hiebler, Herbert; Bernhard, Christian*

**Mechanical Properties and Crack Susceptibility of Steel during Solidification**

*steel research 69 (1998) No. *, p. ??/??*

Hot tensile tests on steel shells during solidification. Description of test method. Characterisation of cooling conditions. Evaluation of structure parameters. Influence of carbon content on high temperature strength and crack susceptibility. Influence of solidification structure and structure fineness.

**Mechanische Eigenschaften und Rißempfindlichkeit von Stahl während der Erstarrung**

Heißzugversuche an Stahl-Strangschalen während der Erstarrung. Beschreibung der Versuchsanordnung. Charakterisierung der Strukturparameter. Einfluß des Kohlenstoffgehalts auf Hochtemperaturfestigkeit und Rißempfindlichkeit. Einfluß der Erstarrungsstruktur und Strukturfeinheit.

# Mechanical Properties and Crack Susceptibility of Steel during Solidification

**Herbert Hiebler and Christian Bernhard**

The determination of high temperature mechanical properties is a necessity for the better understanding of defect formation during continuous casting of steel. Although a lot of studies have been performed on this topic, some aspects like the influence of cast structure on crack susceptibility are still unclear. The present paper focuses on the effect of carbon on high temperature strength and crack susceptibility. Tensile tests have been performed on steel shells with different carbon content during solidification. The results indicate an important influence of the orientation of the inhomogeneous cast structure towards main stress axis. The initial shell with a thickness of only a few millimetres is more sensitive to defect formation than commonly believed. The critical limits of straining range from about 0.15 to 0.4 %. With increasing shell thickness, the critical strain ascends up to 1.6 %. The detected segregated internal cracks form within the critical temperature range. The strength near solidus temperature is lower than in comparable hot tensile tests, which can also be attributed to the detrimental effect of the inhomogeneous structure. The strength at solidus temperature amounts to 2.5 MPa for higher carbon steels and about 1 MPa for low carbon steels.

**Mechanische Eigenschaften und Rißanfälligkeit von Stahl während der Erstarrung.** Die Ermittlung von mechanischen Hochtemperatur-kennwerten ist eine Voraussetzung für das bessere Verständnis der Bildung von Rissen während des Stranggießens von Stahl. Obwohl eine Vielzahl von Untersuchungen zu diesem Thema vorliegen, bleiben Fragen, wie jene nach dem Einfluß der Gußstruktur auf die mechanischen Eigenschaften unklar. Die vorliegende Arbeit behandelt den Einfluß des Kohlenstoffs auf die Hochtemperaturfestigkeit und Rißanfälligkeit von Stahl. Es wurden Zugversuche an erstarrenden Strangschalen mit unterschiedlichen Kohlenstoffgehalten durchgeführt. Die Resultate deuten auf einen deutlichen Einfluß der Orientierung zwischen Dendritenwachstum und Beanspruchungsrichtung. Während der Anfangserstarrung ist die Stahlschale wesentlich empfindlicher als allgemein angenommen. Die kritischen Grenzwerte für die aufgebrachte Gesamtdehnung liegen bei 0.15 bis 0.4 %.Mit zunehmender Schalendicke steigt die kritische Dehnung auf 1.6 %. Die detektierten geseigerten Innenrisse werden im kritischen Temperaturintervall gebildet. Die Festigkeit nahe der Solidustemperatur ist niedriger als bei vergleichbaren Heißzugversuchen, was auf den schädlichen Einfluß der inhomogenen Erstarrungsstruktur zurückzuführen ist. Die Festigkeit bei Solidustemperatur liegt bei rund 2.5 MPa für höhergekohlte Stähle und 1 MPa für niedriggekohlte Stähle.

## Introduction

The prevention of defect formation during continuous casting demands both, detailed knowledge of high temperature mechanical properties of the cast material, and analysis of the strand deformation during the continuous casting process. Innumerable studies have shown, that high temperature ductility (HTD) of steel is characterised by more or less strongly developed, temperature dependent zones of embrittlement. The results of these studies, with priority to the influence of steel composition on crack susceptibility have recently been compiled by K. Schwerdtfeger [1].

The present work concerns the embrittlement above solidus temperature, the so-called “Brittle Temperature Range I” (BTR I), and the influence of carbon content on high temperature strength just below solidus.

The first increase of ductility during solidification occurs with complete solidification, and thus, the Zero-Ductility-Temperature (ZDT) is often equated with solidus temperature (solid fraction between 0.98 and 1) [2-5]. As opposed to this, the ramification of the secondary dendrite arms and capillary forces of the last residual liquid between the dendrites enable the solidifying material to transmit forces below Zero-Strength-Temperature (ZST) according to solid fractions of around 0.65-0.8 [2, 4]. The temperature range between ZST and ZDT is the so-called Critical Temperature Range (CTR) for internal crack formation [5]. Overcritical deformation perpendicular to dendrite growth direction leads to dendrite separation, preferably along primary grain boundaries. The growing interstices are filled with enriched melt and remain as segregated streaks

in the product, even after rolling or forging with high deformation degrees.

According to common concepts, increasing CTR points to higher danger of inner crack formation during casting. Therefore, the determination of ZST and ZDT in hot tensile tests allows a qualitative indication of crack susceptibility of individual steel grades. In general, heavy segregating elements, like sulphur or phosphorus, lower ZDT, widen CTR and increase the hazard of internal crack formation [1].

Whereas the influence of steel composition on crack susceptibility in BTR I seems - at least qualitatively - sufficiently examined, the influence of the cast structure is rather unclear. Tensile tests on columnar solidified structures yield distinctly lower ZDT values than tests on equiaxed solidified specimen [6]. This indicates higher crack susceptibility of columnar dendritic structures, according to the results of plant observations [8]. The consideration of the crack formation mechanism makes also clear, that straining of a cast structure perpendicular to columnar dendrites yields poorer mechanical properties than tests parallel to dendrite growth direction. This assumption has not been confirmed by tensile and creep tests if the specimen has been machined from different positions in continuously cast slab, cooled down to room temperature and reheated up to test temperature [6, 7], because the original grain structure has been destroyed by the $\gamma$-$\alpha$ and $\alpha$-$\gamma$ transformation. “In-situ” tensile tests on austenitic stainless steel at varying cooling conditions and radial cooling of the specimen by Argon have shown, that high temperature strength near solidus temperature decreases with increasing part of radial solidified in shell cross section [9]. Nevertheless, isothermal tensile testing above ZDT with stress axis normal to dendrite growth axis seems impossible in conventional tensile testing, as temperature gradients inside the specimen are unavoidable.

Besides morphology and orientation of the solidification structure, the influence of its fineness, characterised by primary dendrite spacing (PDS) and grain size should be considered, too. Whereas the detrimental effect of increasing grain size on crack susceptibility in BTR II (below 1200 °C) is well-known [10, 11], no internal crack criterion includes structure parameters. Only a new model for the prediction of hot tearing in Al-alloys considers the secondary arm spacing (SDAS) of the columnar dendrites [12]. If the fineness of the cast structure has an influence on crack susceptibility - which seems reasonable - the initial nucleation and cooling conditions become also important for the simulation of internal crack formation in the continuous casting of steel.

This is one reason, that beside common hot tensile testing methods some new laboratory scale experiments have been developed. The test principles and results have been recently compiled by M. Wolf [13]. One of these methods is the so-called “Submerged Split-Chill Tensile”(SSCT)-test. Initially developed for hot tensile tests on Al-alloys [14, 15], the method has been adapted for steels, and gains a new insight into the formation of internal cracks and the influence of cast structure on high temperature mechanical properties.

## Experimental

**Testing: Figure 1** presents a schematic view of the SSCT-test method [16,19,24,27,28]. A solid steel test body, split in two halves, is submerged into the liquid melt in an induction furnace. During a holding time, varying from 6 to 14 seconds, a steel shell solidifies around the test body. The cooling conditions are similar to those in continuous casting. This can be either verified by thermal analysis and the calculation of heat flux density on the chill surface [16], or by measuring primary dendrite spacing (PDS) or secondary dendrite arm spacing (SDAS). The cooling conditions can be varied by coating the test body with a thin refractory material layer.

At the end of holding time, the lower half of the test body moves downward via hydraulic force. Force and elongation are recorded. The solidifying shell is subjected to tension perpendicular to main dendrite growth axis, in correspondence with shell deformation in continuous casting at controlled strain rate between $10^{-3}$ and $10^{-2}$ /s.

After tensile testing the test body and the solidified shell emerge immediately out of the melt.

**Metallography:** The metallographic examination of the solidified shell allows the detection of cracks and the characterization of microstructure by means of PDS ($\lambda_1$) and SDAS ($\lambda_2$). **Figure 2** gives an example of PDS and SDAS for a 1% C roller bearing steel SSCT-shell (not coated steel chill, maximum cooling rate), in comparison with a billet caster breakout strand shell (0.62% C, 115x115 mm, casting speed 2 m/min) [17]. The SSCT-test series (later on discussed as group D) has been performed with the aim to simulate billet casting conditions. As the initial local cooling rate is greater 10 K/s, and both steels solidify as austenite phase, the DAS-difference due to the different C-content is small. The good agreement of SDS and PDAS indicates, that the cooling conditions correspond well. The PDS/SDAS-ratio is approximately 3, according to literature [18].

Coating of the test body lowers heat transfer down to slab casting conditions [19] and increases PDS and SDAS. Thus, the cooling conditions during most continuous casting and even near-net-shape casting processes can be simulated.

**Thermal analysis:** The interpretation of the test results requires detailed knowledge of the temperature distribution inside the solidifying shell and shell growth during solidification. Therefore, the increase of temperature inside the test body is recorded in defined distances from the chill-shell interface by means of thermocouples. This enables the calculation of heat flux density at the chill surface [16]. By solving one dimensional heat conduction in cylindrical coordinates (equation 1) the enthalpy distribution between chill surface and the inner side of the induction furnace is determined:

$$\rho \cdot \frac{\delta H}{\delta t} = \frac{1}{r} \cdot \frac{\delta}{\delta r} \cdot \left[ r \cdot \kappa \cdot \frac{\delta T}{\delta r} \right] \quad (1)$$

where ρ is the density, H the enthalpy and κ the thermal conductivity of steel as a function of temperature.
With the initial temperature of the steel bath $T_M$, the radius of the steel chill $r_i$, the inner radius of the induction furnace $r_o$, and the heat flux density at the chill surface q, the following boundary conditions can be defined

$$T = T_M \, , \, r_i \leq r \leq r_o \, , \, t = 0 \quad (1.1)$$

$$\kappa(\delta T/\delta r) = -q \, , r = r_i \, , \, t > 0 \quad (1.2)$$

$$\kappa(\delta T/\delta r) = 0 \, , \, r = r_o \, , \, t > 0 \quad (1.3)$$

The algorithm includes microsegregation models, thus considering the influence of cooling conditions on solute enrichment and non-equilibrium solidus temperature. The comparison of calculated with measured solidus temperatures for different steel grades indicated that the microsegregation models published by Ohnaka [21] and Kobayashi [22] yield the best results [20]. The actually presented results base on the model from Ohnaka

$$C_L = C_0 \cdot \left(1 - \Gamma \cdot fs\right)^{\frac{k-1}{\Gamma}} \quad (2)$$

$$\Gamma = 1 - \beta \cdot k / \left(1 + \beta\right) \quad (3)$$

$$\beta = 4 \cdot \frac{D_s \cdot t_f}{SDAS^2} \quad (4)$$

with the concentration of solute in liquid $C_L$, the initial concentration $C_0$, the equilibrium partition coefficient k, the diffusion coefficient $D_S$ and the local solidification time $t_f$. The liquidus temperature is calculated with the equations of Kawawa /23/:

$$T_L = 1536 - 80[\%C] - 8[\%Si] - 5[\%Mn] - 34[\%P] - 40[\%S] \quad (5)$$

for primary ferritic solidification (C < 0.5%) and

$$T_L = 1536 - 40 - 60([\%C] - 0.5) - 8[\%Si] - 5[\%Mn] - 34[\%P] - 40[\%S] \quad (6)$$

for primary austenitic solidification (C > 0.5%).

**Figure 3** gives an example for shell growth and temperature distribution during solidification. The characteristic temperatures are: shell „surface“-temperature $T_O$, non-equilibrium solidus temperature $T_S'$, and mean bulk temperature $T_B$. $T_B$ is a simplified instrument for describing

strength-temperature dependence. In order to compare the results of tests on steels with different composition (and widely varying solidus temperature), the homologous temperature $T_B/T_S'$ is often used instead of $T_B$.

**Stress-strain-curves:** During tensile testing, the force-elongation is curve is recorded. The calculation of shell growth during solidification allows to determine a stress-elongation-curve. Shell cross-section during tensile testing refers to shell thickness for solid fraction $f_S = 1$ (with the exception of tests above solidus temperature), since the mushy zone contributes only a small part to total tensile stress. As loading lasts between 2 and 8 s (depending on strain rate and maximum applied strain), the consideration of shell growth during tensile testing is necessary. **Figure 4** gives an example for typical stress-strain curves of higher carbon steel (100Cr6) at different bulk temperature $T_B$. Both curves show initially nearly linear increase of stress with strain. After reaching a maximum, the two curves behave in a different way: in the first case, stress remains nearly constant, pointing at an equilibrium between work hardening and dislocation creep, similar to secondary creep. In the second case, partial crack formation is the only explanation for the steep decrease of stress. Therefore the appearance of the stress-strain-curves gives a first indication of crack susceptibility of a steel grade under the testing conditions.

The characteristic values out of the SSCT-test curves are $\sigma_p$ (peak stress), corresponding to maximum strength and $\varepsilon_p$, the strain when achieving $\sigma_p$. The maximum applied strain during testing exceeds 1.2 % in most cases. This is the order of magnitude of common critical strain values [13]. The temperature dependence of $\sigma_p$ is given by an empirical equation [24]:

$$\sigma_p = \sigma_m \cdot e^{\beta \cdot \left(1 - \frac{T_B}{T_S}\right)} \qquad (7)$$

$\sigma_m$ is the strength at solidus temperature and $\beta$ describes the influence of temperature on strength. For every tested steel grade, $\sigma_m$ and $\beta$ are determined. As can be seen later on, $\sigma_m$ is a helpful tool to describe the effect of alloying elements on high temperature strength.

## Results and Discussion

**Influence of carbon on high temperature strength near solidus temperature: Table 1** gives the composition of four different groups of carbon steels, which have been tested. Group A are low carbon, primary ferritic solidifying steels. Group B are steels with carbon contents ranging from 0.17 to 0.31 %, and thus undergoing the peritectic reaction during solidification. Group C and D are high carbon, primary austenitic solidifying steels. All tests have been performed at strain rates between $10^{-3}$ and $10^{-2}$ $s^{-1}$.

**Figure 5** presents $\sigma_p$ versus $1 - T_B/T_S'$ and the calculated temperature dependence of $\sigma_p$, according to equation (7). According to the results of hot tensile and creep tests [25, 26], higher carbon steels show higher strength than low carbon steels as they solidify primary austenitic, without phase transformation in the regarded temperature range

(minimum surface temperature $T_O$ around 1060 °C). Steels with less than 0.1 % C transform from δ-ferrite to austenite during cooling at $T_{A4}$. Therefore the part of the cross section with temperature below $T_{A4}$ is austenitic whereas the part at higher temperature is ferritic. This makes it necessary to consider the δ-γ-transformation for an exact determination of high temperature strength, which has been realised with the help of a temperature and phase dependent stress model as reported elsewhere [27]. For a fundamental discussion of the influence of carbon on high temperature strength, the simplified assumption in equation 7 seems sufficient.

As can be seen from figure 5, there is no greater difference between the strength of group C and D near solidus temperature. This points to a dominating influence of the modification of the solidifying material, and the absence of an alloy-hardening effect in this temperature range. This suggests also, that group B should behave in the same manner, because this steel grade is also austenitic at solidus temperature, and free of transformation during cooling in the regarded temperature range. Actually, the strength of steel grade B is distinctly lower at solidus temperature. This may be attributed to a small part to the Boron content (16-71 ppm) of these steels [27, 28], but seems primarily due to structural effects. Steels B undergo the transformation $\delta + L \rightarrow \gamma + L$ before solidification, and according to the results of Revaux and co-workers [10], this leads to coarse austenite grains. Own measurement yields somewhat higher PDS-values for steels of group B in comparison with steels of group C and D, also indicating larger grains [27]. As opposed to this, Lamant and co-authors obtain finer solidification grains near the surface at higher cooling rates for 0.3 % C steel in comparison with 0.6 % C steel, but strongly influenced by convection effects [11]. In any case, the fineness of the solidification structure of steel B is different from that of steel C and D and affects its mechanical properties near solidus temperature. A well-founded interpretation of the results demands further work into the characterisation of the micro- and macrostructure immediately after solidification. As already pointed out elsewhere [1], this will be an essential necessity for a better understanding of the relationship between microstructure and high temperature mechanical properties.

**Table 2** compiles the evaluated $\sigma_m$ and β-values for the different steel grades, according to equation (7). The strength at solidus temperature is, especially for the higher carbon steels, significantly lower than values from literature. Published strength values at solidus range from 4 to 8 MPa for austenite [2,4,9], compared with $\sigma_m$ = 2.4 MPa for group C and D. Ferritic iron, respectively low carbon steels, yield strength values between 1 and 2 MPa at the melting point [4,29,30], thus corresponding with the evaluated $\sigma_m$-values of group A (1 MPa).

The difference of the strength seems due to the influence of the orientation between dendrites and main stress axis. This would also explain, that higher carbon steels with distinct microsegregations and grain boundaries segregations are more sensitive to normal oriented straining, than low carbon steels. This may not be observed, if the specimen is cooled down to room temperature and reheated

before tensile testing, as phase transformation, precipitation and the reduction of enrichment by diffusion lower the detrimental effect.

**Strain-rate dependence of strength:** The maximum stress or the first peak stress in a hot tensile test is usually related to the strain rate by an empirical equation in the form:

$$\sigma_p = A \cdot \dot{\varepsilon}^{n} \qquad (8)$$

with A as a function of temperature, and n as a constant.

Even though the strain rate varies only in the small range within $10^{-3}$ and $10^{-2}$ /s in order to simulate continuous casting conditions, the determination of n in equation (8) gives an indication of the validity of material laws for SSCT-test results. The analysis of the $\sigma_p$-values versus temperature and strain rate for steels B yields a strain rate hardening exponent n of 0.22. This corresponds with the expected value of 0.2–0.3 for steel according to a compilation of creep data by Harste and Schwerdtfeger [25, 26].

**Influence of carbon content on internal crack formation:** as pointed out in the description of the test method, the metallographic examination of the solidified shell allows the detection of defects. **Figure 6** gives an example for the microstructure of a primary etched (Bèchet-Beaujard) shell cross section (0.6 % C, 0.7 % Mn, S and P < 0.01%), with a segregated internal crack. The tensile test has been performed at a strain rate of $10^{-2}$ /s, the applied strain amounts to 4 %. Even though 0.6 % C steels with high Mn/S-ratio (≈120) and low P-content are rather crack insensitive, this deformation exceeds the critical limit. The bottom of the crack can be found in a distance of about 5.2 mm from the chill-shell interface, and has a length of 2.5 mm. The comparison with the solidification calculation shows, that the crack occurs between the isotherms for solid fraction $f_s$ = 0.8 and 1.0, within the critical temperature range (CTR). This applies to all tested steel grades and confirms the above-mentioned crack formation concept.

In contrast with isothermal tensile testing, CTR moves during the loading period, and the SSCT-method thus also considers the influence of the shell growth rate on crack propagation. Increasing cooling rate yields higher solidification velocity and a shorter dwell time within the CTR. This will reduce the extent of crack growth, and especially at low strain rate, the cracks may be “overgrown” by the liquid/solid interface [13, 24].

As can also be seen in figure 6, internal cracks occur often along primary grain boundaries. This is another reason, that coarse grains worsen the crack sensitivity, not only within BTR II, but also during solidification.

In order to quantify the defect formation for every test, four etched micro-sections are examined under the optical microscope. The number of detected defects, multiplied with an empirical significance-factor (1 for segregated internal cracks and surface cracks, 2 for open internal cracks, and 3 for total breaks), are summarised in a defect index. **Figure 7** presents the defect index versus homologous temperature for the tests of group C. The open squares represent tests with a maximum applied strain of

1.6 %, the solid squares such ones with a higher strain up to 4 %. As can be seen, thin shells with higher bulk temperature are very crack sensitive, and straining by 1.6 % exceeds the critical limit of shell deformation in any case. The main reason for the high crack susceptibility is uneven shell growth, which leads to strain and stress concentration in weak points. After crack nucleation, the cracks propagate very fast and often cause breaks of the shell and the penetration of liquid steel. The corresponding stress-strain curves are characterised by strongly developed peaks, followed by a steep decrease of stress, according to crack propagation and the decreasing shell cross section. $\varepsilon_p$, the strain when achieving $\sigma_{p,}$ offers a useful strain criterion during initial solidification in the near-meniscus zone of a continuous casting mould. The determined $\varepsilon_p$-values range within 0.15 and 0.35 %. These are of course mean values over the height of the shell specimen, as the maximum strain in weak points is much higher. But the formation of an uneven shell and following strain concentration is also typical for the initial solidification in continuous casting. This is another reason why the testing conditions match well with the condition in a continuous casting mould.

With increasing shell thickness and lower bulk temperature, crack susceptibility decreases significantly. Shells with a mean thickness of around 10 mm, corresponding with a continuous casting shell at the exit of a mould, resist to strains of 1.6 % and more. This stands in good agreement with published results of Yamanaka and co-authors [5]. Obviously, the part of the shell at lower temperature and higher strength prevents uneven strain distribution and the formation of weak points over the height of the specimen. This outweighs the detrimental effect of a lower solidification velocity and the resulting increase of dwell time within CTR, which is responsible for a decrease of critical strain with solidification time in continuous casting [31].

The tendency towards higher critical strain values with increasing shell thickness can be obtained for all tested steel grades. During initial solidification even crack insensitive steels, like that ones of group A show critical strain of 0.2 % and about 1.6 % when the shell thickness increases up to 10 mm. Thus, the results depend not on the critical temperature range which is much wider for the higher carbon steels.

As already mentioned, the form of the stress-strain curves gives a good indication of crack appearance. **Figure 8** presents the defect index versus $\varepsilon_p$. The open squares and triangles symbolise tests with stress peaks and subsequent steep decrease, leading to defect formation in every case. If stress remains at a more or less constant level after reaching the maximum stress, no or only a few defects form. The maximum applied strain ($\geq$ 1.6 %) exceeds $\varepsilon_p$ for all tests. Obviously, the difference between $\varepsilon_p$ and applied strain determines the number and significance of formed defects. Therefore, the use of $\varepsilon_p$ as critical strain is a safety limit to prevent crack formation.

## Conclusions

The SSCT-test method has proved as a useful tool for the determination of high temperature mechanical properties of steel during solidification under continuous casting conditions

As expected from the above-mentioned concepts for the influence of solidification structure on high temperature plasticity, the determined strength values are lower than those from conventional hot tensile tests. Especially higher carbon steels with distinct microsegregations and segregated grain boundaries yield strength values at solidus temperature which amount only to about the half of published data. This points to the importance of the orientation of dendrite growth against main stress axis. The fineness of the structure has also some influence on the mechanical properties, but even more on crack nucleation and propagation. Further investigations into the connection between solidification structure and crack frequency will be the next goal of this research, as the previously published results seem rather contradictory.

The metallographic examinations confirm the assumption, that segregated internal cracks form within the critical temperature range, although no significant influence of the width of CTR on crack susceptibility could be found for plain carbon steels. The initial shell is very sensitive to tensile stresses, as uneven shell growth yields stress and strain concentration in weak points. The critical strain of thin shells ranges from 0.2 to 0.4 %. With increasing shell thickness, the critical strain ascends up to 1.6 %. This is in good correspondence with values from literature [5]. The stress-strain curve gives also an information on crack susceptibility. Curves with distinct peak stress and subsequent stress drop indicate defect formation.

In summary, the presented results show, that the initially formed shell is much more sensitive to defect formation as commonly believed. This is the consequence of the inhomogeneous structure of the solidifying material. Only the straining perpendicular to dendrite growth during solidification yields results with relevance to the continuous casting mould. The variation of cooling conditions is a further necessity to investigate the influence of structure on mechanical properties. All these requirements are fulfilled by the SSCT-test method, which gains a new insight into the formation of internal cracks and the influence of cast structure on high temperature mechanical properties.

## Acknowledgement

This paper bases on results of the co-authors doctoral thesis “Mechanical properties and crack susceptibility of solidifying steel under continuous casting conditions”. The authors would like to express their gratitude to Professor Dr.-Ing. K. Schwerdtfeger for serving as external examiner and for his valuable suggestions.

(A 01 ***; received: **. Month Jear)]

*Univ.Prof. Dipl.-Ing. Dr.-mont. Herbert Hiebler; Dipl.-Ing. Dr.-mont. Christian Bernhard, Department of Ferrous Metallurgy, University of Leoben, Austria*

## References


[1] Schwerdtfeger, K.: Crack Susceptibility of Steels in Continuous Casting and Hot Forming, Verlag Stahleisen, Düsseldorf 1994
[2] Nakagawa, T.; T. Umeda; J. Murata; Y. Kamimura and N. Niwa: ISIJ International Vol. 35 (1995), p. 723/729
[3] Kobayashi, S.: Tetsu-to-Hagané Vol. 73 (1987), No. 12, p. S896
[4] Shin, G.; T. Kajitani; T. Suzuki and T. Umeda: Testsu-to-Hagané Vol. 78 (1992), p. 587/593
[5] Yamanaka, A.; K. Nakajima and K. Okamura: Ironmaking and Steelmaking Vol. 22 (1995), p. 508/512
[6] Fujii, H.: Tetsu-to-Hagané Vol. 63 (1977), p. S145
[7] Suzuki, T.; K.-H. Tacke and K. Schwerdtfeger: Met. Trans. Vol. 19A (1988), p. 2857/2859
[8] Fujii, H.; T. Ohashi und K. Hiromoto: Trans. ISIJ Vol. 18 1978, 510-518
[9] Scheller, P.R.; P. Papaiacovou; D.-Y. Lin; M. Cöl and W. Dahl: steel research Vol. 66 (1995), p. 530/536
[10] Revaux, T.; P. Deprez; J.-P. Bricout and J. Oudin: ISIJ International Vol. 34 (1994), No. 6, p. 528/535
[11] Lamant, J.Y; Marchionni, C.; Roch, F. and M. Bobadilla: Solidification des grains de peau en coulée continue, Final Report, EUR 18648, European Commission , Luxembourg 1998
[12] Drezet, J.-M. and M. Rappaz: Proc. Modeling of Casting, Welding and Advanced Solidification Processes VIII, 1998, San Diego, p. 883/890
[13] Wolf, M.: Continuous Casting Volume Nine - Initial Solidification and Strand Surface Quality of Peritectic Steels, The Iron and Steel Society, Warrendale 1997
[14] Ackermann, P.; W. Kurz und W. Heinemann: Mat. Science and Eng. Vol. 75 (1985), p. 79/86
[15] Wagnieres, J.D. und P. Ackermann: La Revue Politechnique Vol. 6 (1985), p. 669/673
[16] Xia, G.; J. Zirngast; H. Hiebler und M. Wolf: Proc. 1st Conference on Continuous Casting of Steel in Developing Countries, The Chinese Society for Metals, Beijing 1993, p. 200/210
[17] Wolf, M.; Clyne, T.W. and W. Kurz: Arch. Eisenhüttenwes., Vol. 53 (1982), p. 91/96
[18] Ohashi, T. and K. Asano: 2nd Japan-Germany Seminar, Tokyo 1976
[19] Hiebler, H. and Bernhard, C.: Publ. Univ. of Miskolc (Hungary), Series B, Metallurgy, Vol. 39, (1995), No. 1, p. 363/372
[20] Bernhard, C. and Hiebler, H.: Proc. Eisen- und Stahlmetallurgisches Kolloquium 1996, Leoben (Austria), p. 98/123
[21] Ohnaka, I.: Trans. ISIJ, Vol. 26 (1986), p. 1045/1051
[22] Kobayashi, S.; T. Nagamichi and K. Gunji.: Trans. ISIJ, Vol. 28 (1988), p. 543/552
[23] Kawakami, K.; Kitagawa, T.; Murakami, K.; Miyashita, Y.; Tsuchida, Y. and Kawawa, T.: Nippon Kokan Technical Report, No. 36 (1982), p. 26/54
[24] Hiebler, H.; J. Zirngast; C. Bernhard und M. Wolf: Proc. Steelmaking Conference Vol. 77, ISS 1994, p. 405/416
[25] Harste, K.; T. Suzuki and K. Schwerdtfeger: Mat. Science and Tech. Vol. 8 (1992), p. 23/33
[26] Harste, K. and K. Schwerdtfeger: Mat. Science and Tech. Vol. 12 (1996), p. 378/384
[27] Bernhard; C.: Mechanical properties and crack susceptibility of solidifying steel under continuous casting conditions, Leoben, 1998, Dr.-thesis
[28] Bernhard, C; Hiebler, H. and M.M. Wolf: Proc. Conf. on Continuous Casting of Steel in Developing Countries, The Chinese Society for Metals, Wuhan 1997, p. 224/229
[29] Wray, P.J.: Met. Trans. Vol. 13A (1982), p. 125/134
[30] Mizukami, H.; M. Kawamoto and T. Watanabe: Proc. Solidification Processing ´97, Sheffield 1997
[31] Kojima, S.; T. Matsukawa and M. Kodama: Proc. Process Technology Conference Vol. 3 (1982), p.255/262

**Figure 1.** Experimental setup, schematic, 1 = lower part of the test body, movable via hydraulic force; 2 = upper part, rigid

**Figure 2.** PDS and SDAS versus distance from chill surface, comparison with data from literature [17]

**Figure 3.** Shell growth and temperature distribution during solidification

**Figure 4.** Typical stress-strain curves with characteristic values

**Figure 5.** Peak stress $\sigma_m$ versus homologous temperature $1\text{-}T_B/T_S$

**Figure 6.** Microstructure of a primary etched (Bèchet-Beaujard) specimen with segregated internal crack along primary grain boundary (top), shell growth during solidification with time scale and isotherms for solid fraction 0, 0.8 and 1.0.

**Figure 7.** Defect index versus homologous temperature $1 - T_B/T_S$, steel grade C

**Figure 8.** Defect index versus peak strain $\varepsilon_p$, test series C and D, separated into different forms of stress-strain curves

**Table 1.** Composition of the tested steel grades

| Series | C [wt.%] | Si [wt.%] | Mn [wt.%] | ∑P,S [wt.%] |
|---|---|---|---|---|
| A | 0.01-0.07 | 0.1-0.4 % | 0.2-0.6 % | < 0.024 |
| B | 0.17-0.31 | 0.1-0.4 % | 0.3-0.8 % | < 0.030 |
| C | 0.54-0.70 | 0.1-0.4% | 0.1-1.1 % | < 0.027 |
| D | 0.96-1.02 | 0.2 | 0.4 | < 0.014 |

**Table 2.** $\sigma_m$ and $\beta$ (equation 7) for series A, B, C and D

| Series | $\sigma_m$ [MPa] | $\beta$ |
|---|---|---|
| A | 0.94 | 13.8 |
| B | 1.15 | 17.0 |
| C and D | 2.40 | 15.0 |

List of abbreviations, explanation and symbols

| Abbreviations, explanation | |
|---|---|
| HTD | High Temperature Ductility |
| BTR I | Brittle Temperature Range I, above solidus temperature, caused by interdendritic enrichment of segregating elements like S, P or B |
| BTR II | Brittle Temperature Range II, below 1200 °C, caused by precipitation and phase transformation along austenite grain boundaries |
| ZDT | Zero Ductility Temperature, °C, first increase of ductility during solidification, corresponding with solid fraction 0.98 – 1.0 |
| ZST | Zero Strength Temperature, °C, first increase of strength during solidification, corresponding with solid fraction 0.65 – 0.80 |
| CTR | Critical Temperature Range, °C, CTR = ZST-ZDT, indicates internal crack susceptibility |
| PDS | Primary Dendrite Spacing, µm |
| SDAS | Secondary Dendrite Arm Spacing, µm |
| List of Symbols for SSCT-Test | |
| $T_S'$ | Non-equilibrium Solidus Temperature, °C |

| | |
|---|---|
| $T_L$ | Liquidus Temperature, °C |
| $T_O$ | Temperatur of the shell at the shell-chill interface, °C, “Surface-Temperature” |
| $T_B$ | Bulk Temperature, °C, $T_B = 0.5 \cdot (T_S' + T_O)$, mean temperature of the shell during testing |
| $\sigma_p$ | Maximum Tensile Strength, MPa |
| $\sigma_m$ | Tensile Strength at Solidus Temperature, MPa |
| $\varepsilon_p$ | Strain when achieving $\sigma_p$ , % |