

A new hot tearing criterion for steel

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ABSTRACT

Modelling of shape casting processes has been very successful in the past two decades where numerous commercial and in-house softwares are available. However, the modelling of continuous casting processes is not yet common, especially for small to mid-sized companies. Even the in-house solutions available in large companies are mainly used for thermal, flow and stress modelling while metallurgical aspects are almost never considered.

Continuous casting simulation has however been extended beyond the macroscopic phenomena. Today, modules for the calculation of segregation, grain structure, primary phase formation and solid-state transformation are available. The goal of this paper is to concentrate on a frequent problem occurring in the production of continuously cast strands: the formation of solidification cracks or hot tears.

In order to predict the occurrence of hot tears in solidifying parts, a hot tearing criterion based on the ability of the interdendritic flow of liquid to compensate for both the solidification shrinkage and the thermally-induced deformation of the growing dendrites, has been recently derived by Rappaz, Drezet and Gremaud and successfully applied to aluminium alloys [1,2]. Based upon a mass balance performed over the liquid and solid phases, this criterion accounts for the deformation of the solid skeleton and for the after-feeding of the interdendritic liquid: it allows for the calculation of the pressure drop in the liquid at the roots of the dendrites. If the pressure falls below a given cavitation pressure, hot tears will initiate and propagate.

The present paper gives a summary of the main features and assumptions of the new hot tearing criterion applied to steel grades. The influence of the carbon content in both ternary systems Fe-C-P and Fe-C-S on the alloy sensitivity to hot cracking is assessed. Finally, implemented in a FEM model of the continuous casting of steel using the package **calco**soft[®], the hot tearing criterion is able to predict both midway cracks due to surface re-heating and centreline cracks.

1. INTRODUCTION

The continuous casting process for steel is schematically represented in Fig. 1a [3]. This process is widely used owing to its inherent advantages of low cost, high yield, flexibility of operation, and ability to achieve a high quality cast product. Yet, during casting, hot tearing or intercolumnar cracking may occur: this represents a major limitation to casting productivity. Fig. 1b shows some cracks revealed by sulphur print. Progress in modelling in recent years has made it possible to calculate with some degree of accuracy the deformations that a solidifying slab experiences as it travels through a continuous caster [4,5]. The three main causes of deformations are bulging between rolls due to the ferrostatic pressure in the caster, misalignment of rollers and (un)bending of the strand. However, one of the long standing problems in continuous casting is the availability of a hot tearing criterion aimed at predicting the occurrence of solidification cracks.

An in-depth review of the different types of internal and surface cracks that can form during the continuous casting of steel has been carried out by Brimacombe and Sorimachi [6]. These authors made clear how the occurrence of hot tears is linked to the existence of a low ductility region for steel above 1340°C and to the stress generated in the solidifying shell due to ferrostatic and roll pressure, bending and straightening operations.

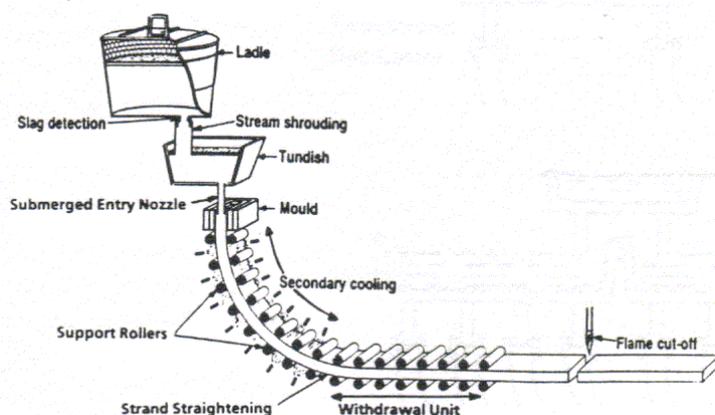


Fig.1a: schematic diagram of a modern slab casting machine showing the main components [3].

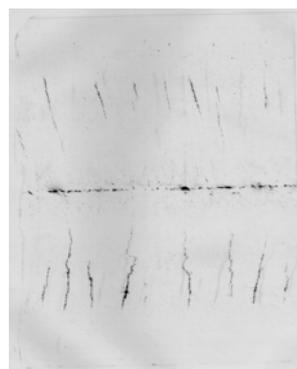


Fig.1b: cracks revealed by sulphur print (courtesy Corus R&D)

Different tests aimed at classifying the alloys with regard to their sensitivity to hot tearing were used in the past, such as the ring mould test [7], the cold finger test [8] and the dog bone test [9] for aluminium alloys, and bending tests performed on solidifying ingots [10] for steel. Nevertheless, the classification of alloys with regards to hot cracking sensitivity still highly depends on the test used.

In opposition to the complexity of the mechanisms involved in hot tear formation, the models developed so far are relatively simple, most of them being based upon the consideration of the solidification interval [11]: the larger the solidification interval of the alloy, the more sensitive it will be to hot tearing. The most sophisticated models use a back-diffusion model [11], for which the maximum hot cracking sensitivity is dictated by the Fourier coefficient in the solid phase. Clyne and Davies [12] have recognised that hot cracking was due to an opening of the mushy zone in a “vulnerable” region where the dendrite arms can be pulled apart easily. They introduced a Cracking Sensitivity Coefficient (CSC) which is given by the ratio of the vulnerability time to the relaxation time. More recently, Won et al. [13] have analysed the

continuous casting process in terms of solidification cracking by introducing the Specific Crack Susceptibility: this parameter represents the averaged possibility of cracking of the strand during the process and is based on the ratio of the stress level to the critical fracture stress. These authors clearly showed that the carbon range sensitive to solidification cracking is between 0.1 and 0.2 wt % C content for an industrial steel grade [13].

In the present contribution, the hot tearing criterion developed by Rappaz, Drezet and Gremaud [1], the so-called RDG criterion, is derived for steel grades. Then, using the microsegregation approach of Clyne, Wolf and Kurz [14] to model the solidification of Fe-C-P ternary alloys, the influence of both C and P content in the hot tearing susceptibility is assessed and compared with common practice. Moreover, the Fe-C-S ternary system is studied with the help of computed solidification paths. Finally, preliminary results obtained with the implementation of the hot tearing criterion in the FEM package **calcosoft**[®] dedicated to continuous casting processes, are presented.

2. APPLICATION OF THE RDG HOT TEARING CRITERION TO STEEL GRADES

A detailed description of the Rappaz Drezet Gremaud model applied to aluminium alloys can be found in [1,2]. In the present section, the derivation of the RDG model for steel grades is carried out. Figure 2 is a schematic diagram of the columnar dendritic growth of steel, similar to what has been observed in transparent organic alloys [1]. In this case, the dendrites are assumed to grow in a given thermal gradient, G , and with a given velocity, v_T . Above a certain volume fraction of solid, mass feeding can no longer compensate for shrinkage, the specific mass of the solid being larger than that of the liquid for most metallic alloys. Therefore, the liquid has to flow from right to left through the dendritic arms.

If the dendritic network is submitted to a tensile deformation, ϵ , perpendicular to the thermal gradient, the flow has also to compensate for that deformation if no hot tears form. The pressure in the interdendritic liquid is schematically represented at the bottom of Figure 2: it decreases from the metallostatic pressure, p_m , at the dendrite tips to a lower value deep in the mushy zone.

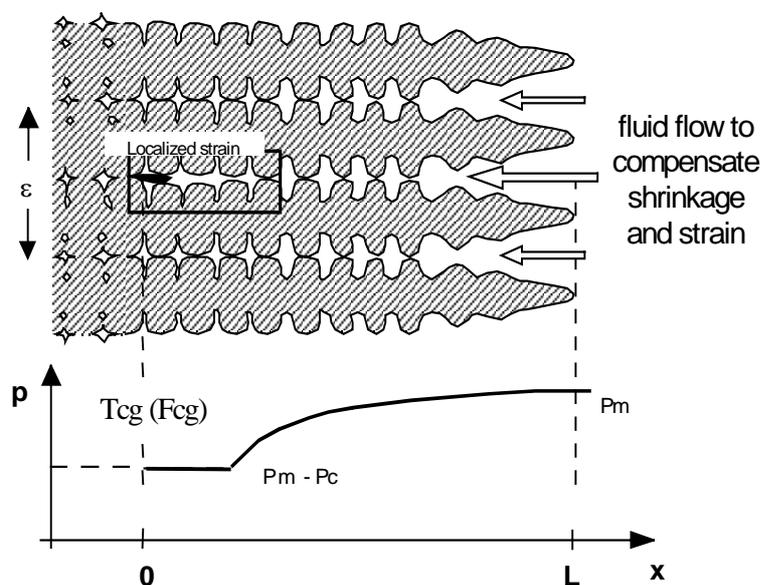


Figure 2: schematic of the formation of a hot tear between columnar grains as a result of a localised strain transmitted by the coherent solid below. The pressure drop in the interdendritic liquid is also indicated.

Above the mass feeding temperature, T_{mf} , the grains have not yet coalesced and the liquid is free to move. On the other hand, below the temperature at which coalescence of the grains takes place, T_{cg} , all the dendrites form a coherent solid network which can transmit the thermal stresses induced by cooling. Note that the temperature at which coalescence between two grains occurs depends on their misorientation and is therefore not unique. Between T_{mf} and T_{cg} , the films of liquid can only resist up to a cavitation pressure at which a void is nucleated and can develop into a hot tear. Any opening of the continuous interdendritic liquid films present in between columnar grains can hardly be compensated for by feeding from the upper region of the mush because of the high volume fraction of solid (i.e., low permeability). The RDG criterion is based on the derivation of the two pressure drop contributions associated with deformation and shrinkage respectively. To do so, a mass balance is performed at the scale of a small volume element of the mushy zone in a reference frame attached to the isotherms [1].

The case of steel is made more complicated by the fact that two phases, δ (ferrite) at low carbon concentration, and γ (austenite) at higher concentration, can form during solidification owing to the presence of a peritectic reaction. Assuming no porosity formation, the sum of the volume fraction of the different phases (liquid, ferrite, austenite) is given by:

$$1 = f_l + f_s = f_l + f_\delta + f_\gamma \quad (1)$$

The specific masses of the three phases, ρ_δ , ρ_γ and ρ_l , are assumed to be constant, but not equal. Two solidification shrinkage factors are then defined:

$$\beta_\gamma = \frac{\rho_\gamma - \rho_l}{\rho_l} \quad \text{and} \quad \beta_\delta = \frac{\rho_\delta - \rho_l}{\rho_l} \quad (2)$$

β_γ and β_δ are respectively around 5.1% and 3.6% for steels [15]. The velocity of the liquid is related to the pressure gradient in the liquid via the Darcy equation. The permeability of the mushy zone is given by the Carman-Kozeny approximation [1]:

$$K = \frac{\lambda^2}{180} \frac{(1-f_s)^3}{f_s^2} \quad (3)$$

where λ is the distance characterising the tortuosity of the mushy medium, typically the secondary dendrite arm spacing. Considering that the fluid moves along the thermal gradient only, whereas both solid phases, δ and γ , deform identically in the transverse direction, one can calculate the pressure within the mush:

$$p = p_a + \rho gh - \Delta p_{sh} - \Delta p_{mec} \quad (4)$$

where p_a is the atmospheric pressure, ρgh the metallostatic contribution. Δp_{sh} and Δp_{mec} are the pressure drop contributions in the mush associated with the solidification shrinkage and the deformation, respectively. In steady state conditions and assuming a uniform thermal gradient, G , throughout the mush, the shrinkage contribution is given by:

$$\Delta p_{sh} = \frac{180\mu v_T}{\lambda^2 G} (\beta_\gamma A_\gamma + \beta_\delta A_\delta) = A \frac{180\mu v_T}{\lambda^2 G} \quad \text{with} \quad A_i = \int_{T_{cg}}^{T_{liq}} \frac{(f_i^0 - f_i) f_s^2}{(1-f_s)^3} dT \quad (i = \delta \text{ or } \gamma) \quad (5)$$

where μ is the viscosity of the liquid phase and f_i^0 is the volume fraction of ferrite ($i = \delta$) or austenite ($i = \gamma$) at the end of solidification. Note that the sum of these two fractions must equal 1.

Moreover, assuming a uniform mechanical deformation rate throughout the mush, $\dot{\epsilon}$, the mechanical contribution is then given by:

$$\Delta p_{\text{mec}} = \frac{180\mu\dot{\epsilon}}{\lambda^2 G^2} \left[(1+\beta_{\delta})B_{\delta} + (1+\beta_{\gamma})B_{\gamma} \right] = B \frac{180\mu\dot{\epsilon}}{\lambda^2 G^2}$$

$$\text{with } B_i = \int_{T_{\text{cg}}}^{T_{\text{liq}}} \frac{f_s^2 \left[\int_{T_{\text{cg}}}^T f_i dT \right]}{(1-f_s)^3} dT \quad (i = \delta \text{ or } \gamma) \quad (6)$$

The coalescence temperature T_{cg} corresponds typically to a solid fraction of 99% between two different grains and is slightly lower, around 95 % for two dendrites of the same grain [15]. Equation 5 reveals that the shrinkage contribution is proportional to the shrinkage factors and to the speed of the isotherms whereas the mechanical contribution is proportional to the strain rate. Both contributions are inversely proportional to the square of the secondary dendrite arm spacing, λ .

The four parameters A_i and B_i depend only on the composition of the steel grade and on its solidification path, i.e. on the relationship between f_s , f_{δ} , f_{γ} and T . They can be calculated using a back diffusion model such as that of Clyne, Wolf and Kurz [14] or more sophisticated numerical microsegregation models.

Eventually, if the pressure, p , given by equation (4), falls below the cavitation pressure, p_c , a hot tear forms. This condition allows to calculate the maximum strain rate sustainable by the mushy zone, $\dot{\epsilon}^{\text{max}}$, and a hot cracking susceptibility, HCS, can be defined as :

$$\text{HCS} = \frac{1}{\dot{\epsilon}^{\text{max}}} \quad (7)$$

The higher HCS, the more susceptible are the alloy and the solidification conditions.

3. INFLUENCE OF C AND P CONTENT ON THE HOT CRACKING SUSCEPTIBILITY

For the sake of simplicity, the ternary system Fe-C-P is considered here. C is known to diffuse relatively well in ferrite and austenite whereas phosphorus, P, like sulphur, S, diffuses slowly. These two elements have a low partition coefficient and are known to produce interdendritic segregate of low melting point [13]. Hereafter, the effect of P is treated in the model. The study of the specific influence of S on solidification cracking will be discussed in the next section.

In order to calculate the integrals A_i and B_i appearing in equations (5) and (6), the solidification path of the alloy is first calculated using the extension of the Clyne and Kurz microsegregation approach for steels [14]. Figure 3 shows the distribution of the liquid, solid, δ and γ phases during solidification of the Fe-0.15 C-0.03P wt %. In this case, it is assumed that no residual ferrite exists after solidification is completed. Using the curves shown in Fig. 3, the integrals A_i and B_i can be computed. The lower limit for the integrals, T_{cg} , or the solid fraction at coalescence, F_{cg} , has a very large influence on the value of the integrals, as reported in Fig. 4 for B_{γ} . Indeed, all integrals tend to infinity when F_{cg} tends to 1. F_{cg} is highly dependent on the concentration of tramp elements, such as P and S, in the residual

interdendritic liquid film. The problem of coalescence has recently been addressed using sharp interface and phase field models [16]. Such models allow to describe the transition between individual grains and a finally coalesced/coherent mush. The amount of bridging, i.e., of necks linking two grains, can be determined with such approaches.

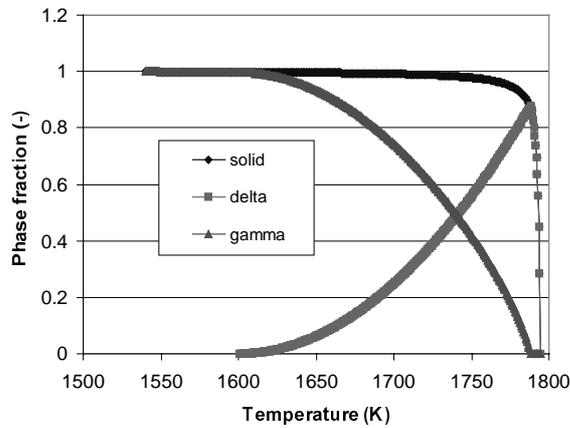


Fig. 3: solidification path of the Fe-0.15 C-0.03P wt % alloy.

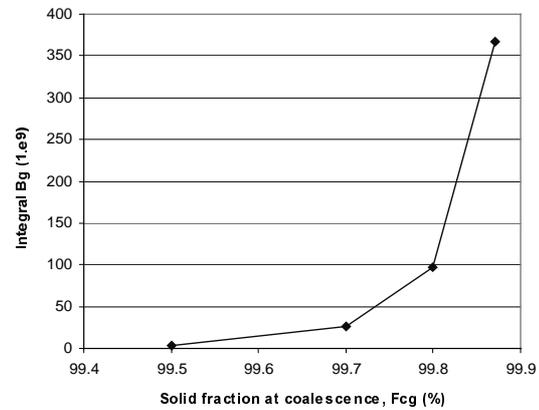


Fig. 4: influence of F_{cg} on B_{γ} for the Fe-0.15 C-0.03P wt % alloy.

Integrals A_i and B_i allow to classify the alloys with regards to the hot cracking sensitivity: indeed for given thermo-mechanical conditions, the larger these quantities, the larger the pressure drop given in Eq. (4). Fig. 5 and 6 compare the results obtained with the present hot tearing model (integral B_{γ}) with the measured index of sub surface cracking, ISSC, reported in [14] as a function of the C content. For C content lower than 0.1 wt %, the alloy solidifies only in the δ structure, whereas at higher C content, solidification terminates with the γ structure. The sudden increase around 0.1 wt % in the hot cracking sensitivity is well reproduced by the present model.

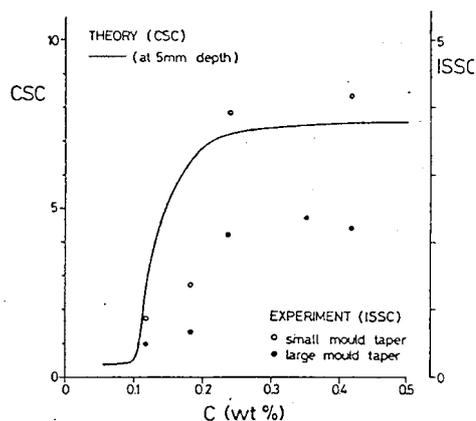


Fig. 5: ISSC as a function of C content [13]

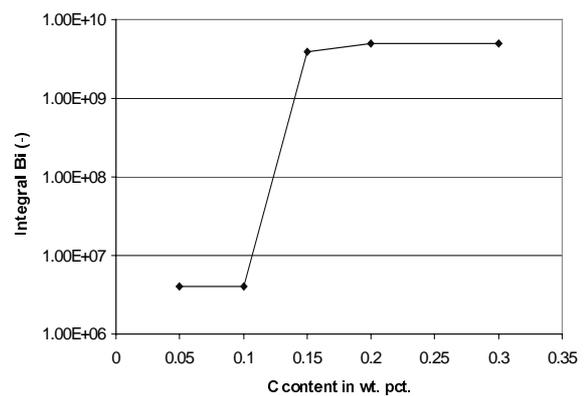


Fig. 6: Integral B_{γ} as a function of C content.

Finally, Fig. 7 shows the dependence of the integral B_{γ} with the P content for two values of F_{cg} , the iron level being kept at 0.15 wt % (the other three integrals behave in a similar way). The two curves exhibit the Λ shape, well known in aluminium binary systems [1,12].

Increasing F_{cg} pushes the maximum towards lower C content. In common steel grades, the P level is kept below 0.03 wt %: Fig. 7 shows that a slight increase in the P content leads to a much higher hot tearing sensitivity.

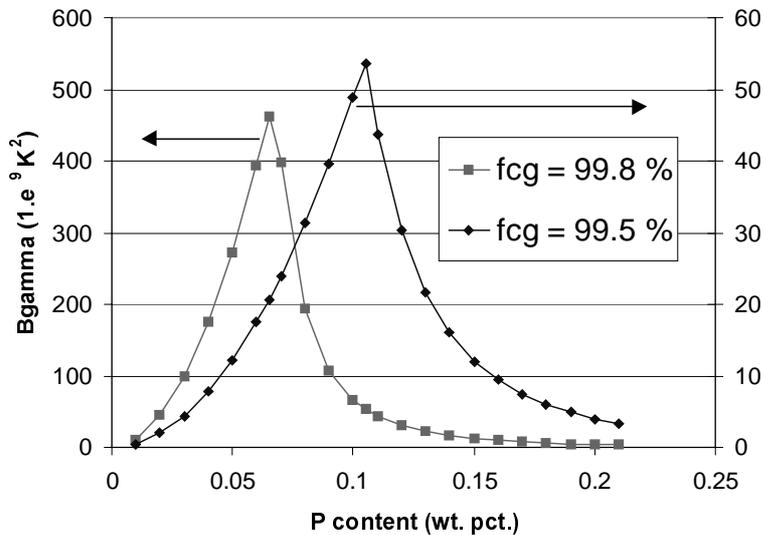


Figure 7: integral B_γ as a function of P content for two values of F_{cg} (Fe content is 0.15 wt. %).

4. INFLUENCE OF C AND S CONTENT ON THE HOT CRACKING SUSCEPTIBILITY

Let us consider now the ternary system Fe-C-S is considered here. In order to calculate the integrals A and B appearing in equations (5) and (6), the solidification paths of the alloys are now computed with the microsegregation numerical model used at Corus R&D [18]. Figure 8 shows the computed distribution of the liquid, solid, δ -ferrite and γ -austenite phases during solidification of a Fe-0.14 C-0.01S wt %. In this case, it appears that some residual ferrite, about 30%, remains after solidification is completed. Using this type of curves, the integrals A and B are computed for a given solid fraction at coalescence, F_{cg} , of 99 %. The results are shown in Figure 9 for a S content of 0.01 wt. % and for no S at all.

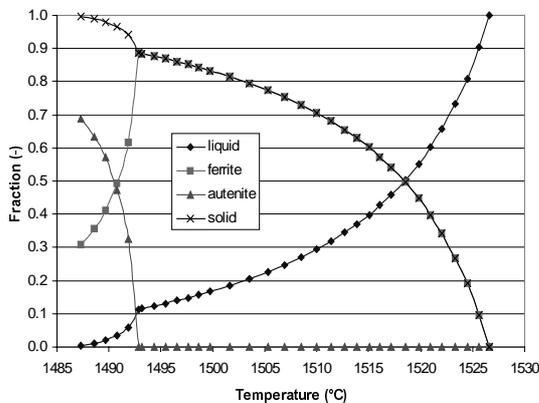


Fig. 8: computed solidification path of the Fe-0.14 C-0.01S wt % alloy.

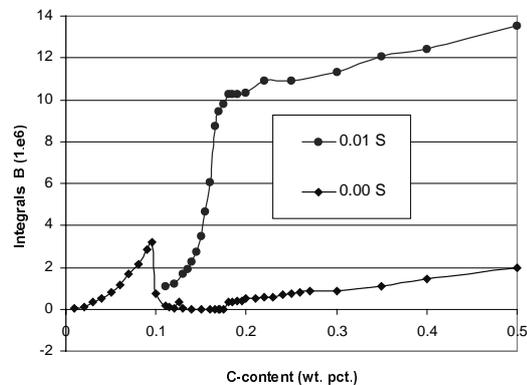


Fig. 9: influence of C content on integral B for two S levels.

In the absence of S, the Fe-C binary system exhibits a Λ shape curve already mentioned in section 3. This behaviour is very similar to the findings of Won et al. [13]. On the other hand, the presence of minute amount of S 0.01 wt. %, drastically influences the overall behaviour. Indeed, the parameter B now rapidly increases around 0.15 wt. % C and continues to increase but much slower at higher C contents. This trend is in accordance with plant observations [18,19].

5. APPLICATION TO THE CONTINUOUS CASTING OF STEEL

The present section shows the results obtained with the implementation of the RDG approach in a FEM thermal model of the continuous casting process built in **calcosoft**[®] [20]. Fig. 10 shows a 3D view of the temperature field obtained in steady state conditions with the help of a space-dependent transport velocity. The temperature and solid fraction fields are presented in Fig. 11 in the mid-plane of the strand. Just after the straightening point, surface re-heating of the slab has been modelled by decreasing the local heat transfer coefficient: the temperature increase is visible at the position of the two arrows on the thermal field of Fig. 11.



Fig. 10: thermal field during CC of steel (3D calculation, half of the strand)

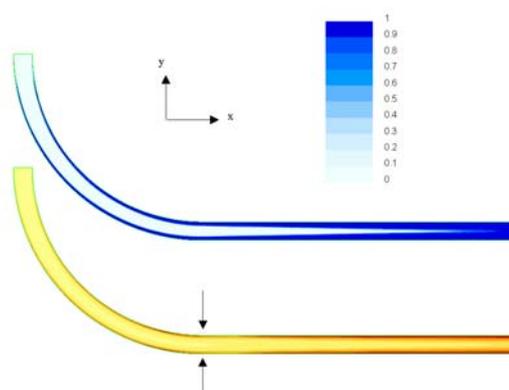


Fig. 11: temperature and solid fraction fields during CC of steel (2D calculation, mid-plane)

Eq. (7) has been implemented in the thermal model and the distribution of HCS along the cast length is reported in Fig. 12. HCS increases as solidification proceeds inside the slab. Nevertheless, a localised maximum is found just after the straightening point where surface reheating was simulated: this maximum is to be related with the occurrence of mid-way cracks, as reported in [6]. Moreover, the largest value of HCS found at the bottom of the liquid sump, where the two solidifying shells impinge, can be related to centreline cracks [6].

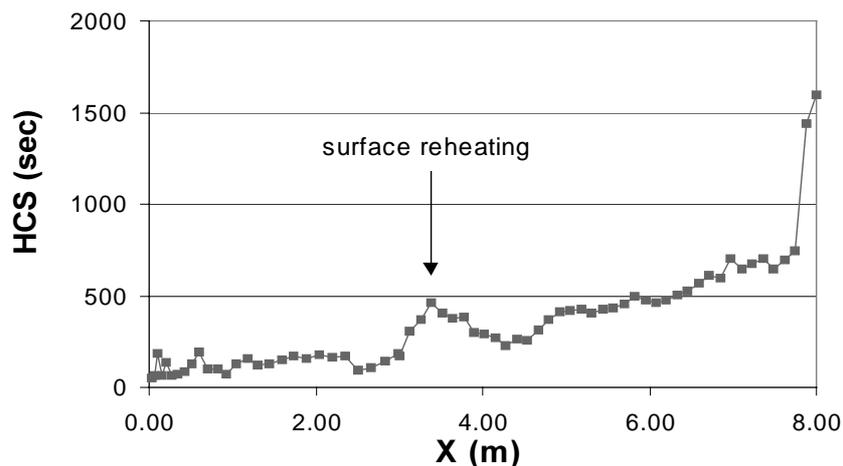


Fig. 12: Hot cracking susceptibility (HCS) as a function of the cast length, X.

CONCLUSION

The Rappaz-Drezet-Gremaud hot tearing criterion, a two-phase model in which both deformation of the coherent solid and interdendritic fluid flow are considered, has been derived for steel grades. The influence of the carbon, phosphorus and sulphur content agrees well with what is reported in practice. In particular, the tremendous effect of the presence of sulphur on the hot cracking susceptibility is well reproduced. One prerequisite to the application of the present approach is to get a good knowledge of the solidification path of the alloy, especially at low liquid fractions and under solidification conditions close to those encountered in the plants. If this condition is fulfilled, cracking sensitivity numbers that include the chemistry of the steel grades [19], could be computed and compared to plant observations.

Moreover, when implemented in a thermal model of the continuous casting of steel, the criterion is able to predict both mid-way and centreline cracks. In order to see the effects of the deformations undergone by the slab, such as bulging between rolls due to the ferrostatic pressure in the caster, misalignment of rollers and (un)bending of the strand, the present criterion needs to be implemented in thermo-mechanical models of the continuous casting process.

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REFERENCES

1. M. Rappaz, J.-M. Drezet and M. Gremaud : "A new hot tearing criterion", Met. Trans. Vol. 30A, pp-449-455, Feb. 1999.
2. J.-M. Drezet and Michel Rappaz: "Prediction of hot tears in DC cast billets" in Light Metals, Ed. J.-L. Anjier, TMS, 2001, New Orleans, pp. 887-893.

3. W. R. Irving : Continuous Casting of Steel, The Institute of Materials, 1993, London.
4. R. Mostert, R. Haardt and K.-H. Tacke: Optimisation of the straightening process in continuous casting, report EUR 19850 EN of the European Commission, 2001, Brussels.
5. R. Mostert, R. Haardt and K.-H. Tacke: The effect of straightening and bulging on intercolumnar cracking of slabs, in Proceedings of the 84th Steelmaking Conference, vol. 84, p. 177.
6. Brimacombe J.-K. and K. Sorimachi: Crack formation in the continuous casting of steel, Met. Trans., vol. 8B, Sept. 1977, p. 489.
7. J.-M. Drezet and M. Rappaz, Study of Hot Tearing in Aluminum Alloys using the Ring Mold Test, in Modelling of Casting, Welding and Advanced Solidification Processes VIII, B. G. Thomas et al Edts (TMS, Warrendale, PA, San Diego 1998).
8. D. Warrington and D. G. McCartney, Cast Metals 2 (1989) 134.
9. F. Decultieux, PhD work, Ecole des Mines de Paris, CEMEF, France, Nov. 1996.
10. Wintz M., M. Bobadilla and J.-M. Jolivet: Fragilité à la solidification des aciers: influence du carbone, du soufre et du phosphore, La Revue de Métallurgie, CIT, January 1994, p.105.
11. John Campbell, *Castings* (Butterworth-Heinemann, Oxford, 1991).
12. T. W. Clyne and G. J. Davies, *The influence of composition on solidification cracking susceptibility in binary alloy systems*, Brit. Found. 74 (1981) 65. See also : Brit. Found. 68 (1975) 238.
13. Y. M. Wong, H. N. Han, T Yeo and K. H. OH, Analysis of solidification cracking using the specific crack susceptibility, ISIJ International, vol. 40 (2000), no. 2, pp. 129-136.
14. T. W. Clyne, M. Wolf and W. Kurz: The effect of melt composition on solidification cracking of steel, with particular reference to continuous casting, Met. Trans., vol. 13B, p. 259, June 1982.
15. W. Kurz and D. J. Fisher, *Fundamentals of Solidification* (Trans Tech Publication, third edition, Aedermannsdorf, 1989).
16. P.-D. Grasso, J.-M. Drezet and M. Rappaz: "Hot tear formation and coalescence observations in organic alloys" in electronic version of Journal of Metals, JOM-e, TMS, January 2002, <http://www.tms.org/pubs/journals/JOM/0201/Grasso/Grasso-0201.html>
17. M. Rappaz, A. Jacot and W. J. Boettinger: "Last stage solidification of alloys: a theoretical study of dendrite arm and grain coalescence", submitted to Met. Mater. Trans.
18. R. Mostert, Corus Research Development and Technology, Ijmuiden, The Netherlands, private communications and J. Miettinen, "Mathematical Simulation of Interdendritic Solidification of Low-alloyed and Stainless Steels", Acta Polytechnica Scandinavica, Chemical Technology and metallurgy series No. 207 (1992) or J. Miettinen, Metall. Mater. Trans. 28B(1997), pp. 281-291.
19. R. Mosters, G. A. de Toledo, L. Condamin, C. Guyon, N. Lepareur, A. S. Normanton , B. Patrick, E. Sanchez and L. Sancho, Influence of chemistry on intercolumnar and surface cracking sensitivity of steel grades, proceedings of the 4th European Conference on Continuous Casting, Birmingham, UK, October 2002.
20. **calcosoft**[®], user manual, Calcom SA, <http://www.calcom.ch>, CH-1015 Lausanne, Switzerland.