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Microstructure and fracture properties of an ultrahigh carbon steel–mild steel laminated composite

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Abstract

A seven layer steel based (mild steel and ultrahigh carbon steel, UHCS) laminated composite was processed by roll bonding. Impact properties were improved in comparison with the UHCS. Delamination plays an important role by deflecting cracks, absorbing energy and imposing the nucleation of new cracks in the next material layer.

Keywords: Ferrous laminated composites; Roll bonding; Charpy impact testing; Delamination

1. Introduction

Composite laminated materials, consisting of different constitutive materials and alternately separated by discrete interfaces, are an interesting family of materials because they are capable of arresting propagating cracks under impact loading conditions. This effect is related to the presence of interfaces that may delaminate under such conditions and is responsible of the high impact and fracture resistance of the composites, much better than the constitutive material components separately [1–4]. This is due to the extrinsic fracture mechanisms that operate altering crack propagation due to the presence of the layer interfaces. Ductility is also expected to be higher than average of the component ductilities as necking is supposed to be delayed by the constraints imposed to the layer interfaces [5–7]. Regarding this point, the stronger is the interface the higher is the expected ductility. Unfortunately, the stronger the interface, the more difficult is the delamination, thus diminishing the impact properties [4]. A compromise must, therefore, be found in order to fulfill the laminated material service requirements. In this regard, laminated composite materials can provide customizable materials for specific applications.

The processing routes to obtain the laminates are of importance to determine their properties. Different processing routes are available such as press and roll bonding [3,4,7]. These techniques differ in the amount of plastic deformation imposed to the constituent materials and in the amount of fresh interface generated during processing that induces improved bonding. Adequate processing, for instance, via roll bonding, can improve their individual strengths by refining their

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respective microstructures and, thus, improving the strength of the overall laminated composite. The bonding temperature is also an important factor to obtain an adequate bonding, by controlling layer interdiffusion and controlling sharpness of the interface.

In this work, an improvement of the impact properties of an ultrahigh carbon steel, UHCS, was obtained by means of a multilayered material containing UHCS and a mild steel, MS. Microstructures of both steels and their interfaces were examined using scanning electron microscopy (SEM) and impact properties of the composite laminated material were assessed using Charpy impact tests.

2. Materials and experimental procedure

Two different steels were employed as constituent materials of the laminated composite. A low carbon content MS of composition 0.035%C–0.28%Mn–0.03%Si and an UHCS of composition 1.55%C–0.49%Mn–0.05%Si–1.55%Cr–0.17%Al (all percentages in mass) were used as constituent materials. The MS and UHCS materials were in rolled sheet form having 4 and 4.5 mm thickness, respectively.

Squares of dimensions 60×60 mm^2 were cut and machined, for both materials, to present clean and smooth faces. Seven squares were piled up forming the sequence UHCS–MS–$\cdots$–UHCS which were hermetically welded by Tungsten Inert Gas prior to high temperature rolling to avoid oxygen penetration and delamination during processing. A two-high rolling mill of 134 mm roll diameter at a rolling speed of 366 mm/s was utilized. Total thickness of the initial laminate was 29.5 mm. High temperature roll bonding was performed at 650 °C. After 1 h at temperature the material was given six series of three passes of about 5% reduction. Between the different series, the material was introduced again in the furnace at temperature during 5 min. The final thickness was 10.5 mm, thus giving a total reduction of about 3:1.

The initial microstructure of the material constituents and also the microstructure of the laminated composite after roll bonding were observed by means of optical microscopy and SEM. Selected pieces were cut, polished and etched with Nital to reveal the microstructure of both steels and the quality of the interfaces.

Hardness measurements were made with a Vickers indenter under loads of 1.96 and 9.8 N during 15 s.

Two mm V-notched Charpy type testing specimens were mechanized from the constituent materials and the laminated composite to perform 294 J Charpy impact tests.

The specimens were cut parallel to transverse direction and the Charpy sample dimensions of the composite were $10 \times 10 \times 55$ mm^3. Three of them were mechanized with the crack arrester orientation (the notch, in the UHC steel, is perpendicular to the layers in this orientation) and other three with the crack divider orientation (the notch acts equally on all the layers in this orientation). The Charpy sample dimensions of the MS and UHCS materials were $10 \times 3.8 \times 55$ and $10 \times 4.5 \times 55$ mm^3, respectively. Fracture surfaces of the impact tested samples were observed in the SEM.

3. Results and discussion

3.1. Microstructure

Fig. 1a and b show the initial microstructure of the UHCS and MS materials, respectively. The UHCS material presents a very high volume fraction of iron carbides, mostly spheroidized of about 1 μm in diameter; also large pearlitic colonies of about 10 μm are observed. On the contrary, as shown in Fig. 1b, there are scarcely carbides in the iron matrix of the MS steel. Its ferritic grain size, L, is about 50 μm. The UHCS is much harder than the MS, and also less ductile. Vickers hardness at room temperature of UHCS is about 480 whereas it is about 100 that of MS.

As a consequence of processing, the initial microstructure suffered changes in both MS and UHCS material constituents. SEM observations were performed to reveal these changes and also to assess the bonding between the two steels. Fig. 2 shows a SEM micrograph where both constituent

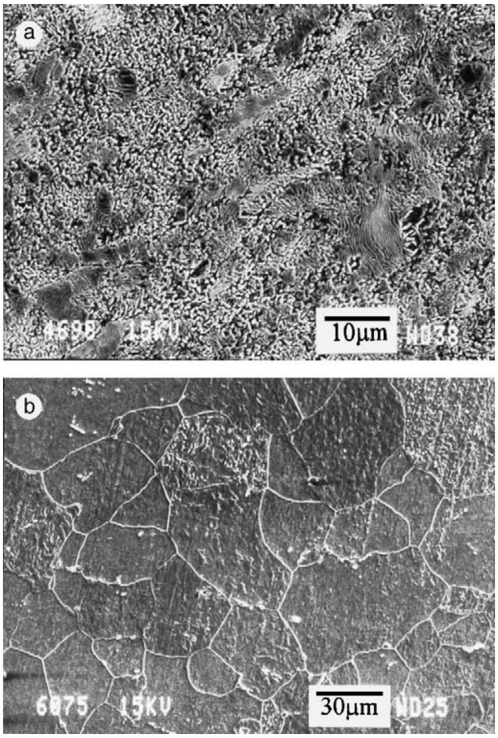

Fig. 1. Initial microstructures observed in the SEM: (a) UHC and (b) MS.

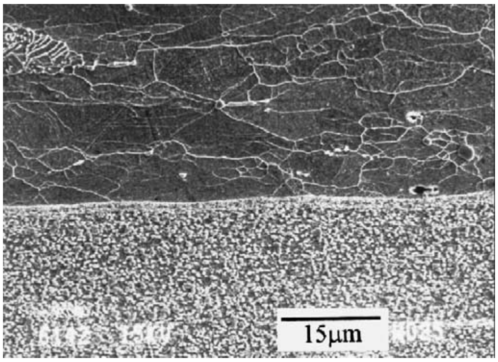

Fig. 2. SEM micrograph of the steel based laminated composite material.

materials and their interface can be observed. The UHCS material presents a refined microstructure due to the rolling reduction of 3:1. There are no visible pearlitic zones and the large carbides have been broken down into small round carbides homogeneously distributed in the ferritic matrix. These small spheroidized carbides are now smaller than 1 μm in diameter. The ferritic grain size is also fine, less than 2 μm. Thus, rolling has provided a refined spheroidized and a homogeneous microstructure of the UHCS. This microstructure grants better mechanical properties than the initial, pearlitic, one, for instance, in relation to ductility and toughness [8,9]. In contrast, the MS shows pancaking of initial grains by rolling where the grain size diminished from 50 μm to about 10 μm in the rolling direction.

The interface between the UHCS and the MS materials is sharp and carbon interdiffusion between layers is not apparent. A band of fine carbides of about 1 μm diameter, however, is observed at the interface. This band is attributed to flattening of the opposing rough surfaces of the constituent materials during rolling, combined with carbon diffusion along this interface eliminating its carbon concentration fluctuations.

3.2. Hardness tests

Fig. 3 shows hardness data of the laminated composite as a function of distance. It is apparent the stacking order of the layers. Those corresponding to UHCS show hardness of about 360, and those corresponding to MS of about 160. Both values are different from those of the initial constituent materials (480 and 100, respectively), due to the microstructural changes occurring during processing. The pearlitic structure of the UHCS is

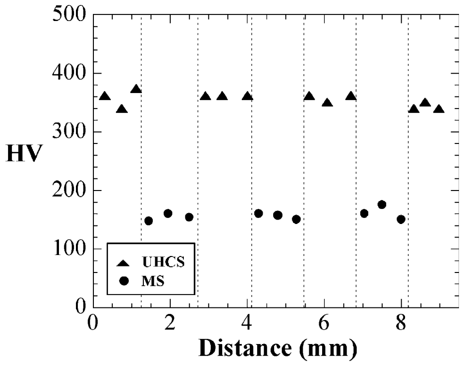

Fig. 3. Hardness Vickers data of the laminated composite material showing the different layers.

destroyed, becoming spheroidized and, thus, decreasing its hardness [10]. On the contrary, extensive grain size refinement and work-hardening of the MS increases its hardness.

3.3. Notch impact tests

Fig. 4 shows Charpy V-notched (CVN) energy data for the monolithic, UHCS and MS materials and also for the composite materials in the crack divider and the crack arrester orientations. A spheroidizing treatment consisting of three temperature cycles of one hour at 740 °C followed by another hour at 650 °C was performed in the monolithic UHCS to produce a microstructure with globular carbides similar to that present in the laminated composite.

The impact values for the laminated composite materials, in either orientation, are much higher than that for the UHCS material. In the most favorable condition, the arrester orientation, an increase of the CVN energy over one order of magnitude is obtained.

On the other hand, the impact value of the laminate composite in the crack arrester direction is slightly higher than that for the MS material. On the contrary, a much lower impact value is found for the divider orientation in the laminate than for the MS material. In fact, the impact value of the laminate in this orientation is lower than that expected from a rule of mixtures, which is 420 kJ/m^2, having into account the volume fractions of the constituent materials. This low energy value in the crack divider orientation can be attributed to the fracture of the laminated composite which is dominated by the fracture of the brittle component, i.e., by the UHCS material. This is in agreement with the failure hypothesis of the weakest link. In the crack divider orientation the crack propagates easily through the thickness of the sample in a defined plane. This indicates that the bonding between layers is not weak and that the UHCS induces brittleness in the MS.

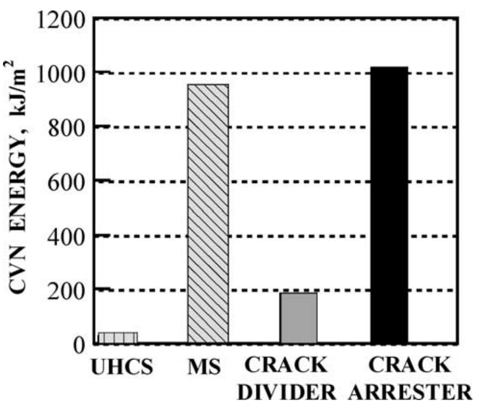

Fig. 4. CVN energy values for the different materials and orientations.

The analyses carried out in the crack divider orientation cannot be applied to the crack arrester orientation since the estimation of the energy values in this orientation is complex because the different layers are not subjected to the same stress state than at the tip of the notch. The possible delamination of the layers transform the notched sample into an unnotched one, implying a large mechanical energy consumption which is difficult to analyze.

3.4. Fractography of impact samples

Fig. 5 shows macrographs of fractured Charpy notched impact samples in the crack arrester and divider orientations. In the arrester orientation, Fig. 5a, the fracture propagates through the interfaces of the various layers of the laminate. The crack propagation is more difficult in this orientation due to the successive discontinuities induced by the interfaces. At these interfaces extrinsic mechanisms of fracture, such as delamination and nucleation of new cracks, take place. An optimum situation would be that for which the interlayer bonding is strong but all the layers delaminate as testing proceeds.

As observed in Fig. 5a delamination occurs more readily between certain layers than between others which is attributed to variations of the bonding strength among layers. This is because delamination and therefore, crack propagation, depend highly on interface bonding. Once delamination takes place at an interface, the crack is stopped at the delaminated surface and a new crack must be nucleated at the next layer. Delamination, therefore, makes difficult the propagation

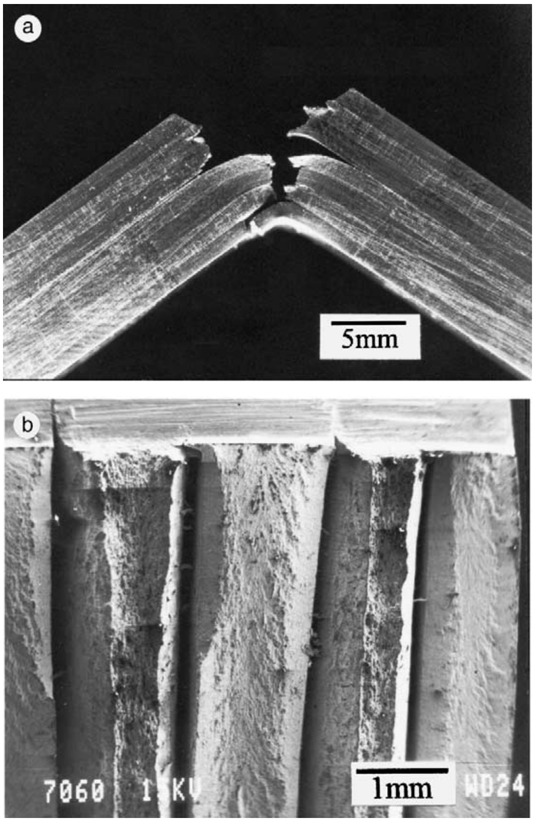

Fig. 5. Fractured Charpy notched impact samples in the orientations (a) crack arrester and (b) crack divider.

of cracks that otherwise would be easy [1,2,4,11–13].

On the other hand, in the divider orientation, delaminations are also observed at every interface, as shown in Fig. 5b. This is indicative that the interface bonding is not excessively strong. Additionally, some necking is observed in all constituent layers, but especially in the MS layers. Thus, there is plasticity of the constituents materials. The reduction of area follows their respective ductilities, i.e., it is larger for the MS than for the UHCS. The higher ductility of the MS accounts for the higher value of impact energy of the composite in the divider orientation with respect to the monolithic UHCS. This is due to the overall higher absorbed energy for plastic deformation during the impact test. An interesting feature also observed in Fig. 5b is that, although the UHCS material absorbs little energy during impact, its fracture micromechanism is ductile type, showing microvoids coalescence. This allows improvements in the fracture and impact properties of the UHCS-containing laminated composites because ductility contributes to the start of delamination through the presence of a necked region.

Fig. 6 shows a detail of fracture surface in the crack arrester orientation, near the notch, where the MS material presents ductile and brittle zones. This is probably due to the fact that this first layer (and each of them successively), is tested mainly in tension deforming and tearing initially in a ductile manner until the cleavage stress is reached and final fracture takes place in a brittle manner. This occurs in the MS but not in the UHCS that remains ductile because the cleavage stress is not reached. It is likely that the more resistant UHCS material induces deformation of the MS at a high stress, thus reaching the MS cleavage stress. The impact and fracture properties could be improved, therefore, by changing the MS steel by a more ductile steel and/or with a higher cleavage stress; for example, a microalloyed steel.

In summary, the fracture and impact properties are importantly influenced by the bonding strength. The latter controls delamination, which, in turn, determines whether crack deflection and re-nucleation takes place during fracture of the laminated composite, absorbing a great amount of energy. To obtain good impact properties in the arrester orientation, which is the important

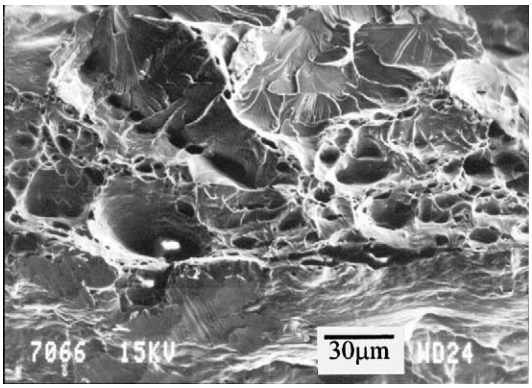

Fig. 6. Detail of a impacted sample showing ductile and brittle zones in the MS close to notch (at the bottom).

orientation for technological applications such as armour-platings, slightly weaker interfaces are preferred to ensure that all layers delaminate. In this configuration, delaminations and successive crack re-nucleations in the next layers maximize absorption of energy. Additionally, the impact properties of the laminated composite are expected to be improved by increasing the number of layers and reducing their thickness.

4. Conclusions

A seven layer laminated composite material, composed of UHCS and MS, was processed by roll bonding. As a consequence of the imposed hot rolling conditions the microstructures of the constituent steels were refined, remaining the interfaces abrupt. Regarding the impact properties it is concluded that:

(1) The Charpy notched impact tests of the laminated composite show better values than those of the UHCS steel, for both crack arrester and crack divider orientations. However, compared to the MS, only improvement in the crack arrester orientation is obtained.
(2) The fracture micromechanism of UHCS is ductile. On the contrary, the final fracture of MS is brittle. It is believed that UHCS induces deformation of the MS at a high stress, thus reaching its cleavage stress.
(3) The crack arrester orientation reveals that extrinsic fracture mechanisms are operating, such as delamination and crack nucleation. Delamination, controlled by interface bonding, plays a key role by deflecting the crack, absorbing energy and imposing the nucleation of new cracks in the next material layer.
(4) The crack divider orientation shows necking, especially in the MS layers, due to plastic deformation during the impact test. However, the impact energy values are worst than predicted by the rule of mixtures, indicating that the bonding between layers is not weak and that the UHCS induces brittleness in the MS.

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References

[1] Kum DW, Oyama T, Wadsworth J, Sherby OD. J Mech Phys Solids 1983;31:173.
[2] Lee S, Wadsworth J, Sherby OD. Res Mechanica 1990;31:233.
[3] Kum DW, Oyama T, Ruano OA, Sherby OD. Metall Trans A 1986;17:1517.
[4] Lesuer DR, Syn CK, Sherby OD, Wadsworth J, Lewandowski JJ, Hunt WH. Int Mater Rev 1996;41:169.
[5] Banu Prakash Babu G, Dube RK. ISIJ Int 1996;36:1184.
[6] Syn CK, Lesuer DR, Sherby OD. Mater Sci Eng A 1996;206:201.
[7] Syn CK, Lesuer DR, Wolfenstine J, Sherby OD. Metall Trans A 1993;24:1647.
[8] Lesuer DR, Syn CK, Goldberg A, Wadsworth J, Sherby OD. JOM 1993;45:40.
[9] Lesuer DR, Syn CK, Sherby OD. Acta Metall Mater 1995;43:3827.
[10] Taleff EM, Syn CK, Lesuer DR, Sherby OD. Metall Mater Trans A 1996;27:111.
[11] Fernández A, Chao J, Carsí M, Peñalba F, Ibáñez J. In: Carsí M, Peñalba F, Ruano OA, Fernández BJ, editors. VII Congreso Nacional de Tratamientos Térmicos y de Superficie, Barcelona, Spain, 59. 1998. p. 65.
[12] Lesuer DR, Syn CK, Riddle R, Sherby OD. In: Lewandowski JJ, Hunt Jr WH, editors. Intrinsic and extrinsic fracture mechanisms in inorganic composite systems. The Minerals, Metals & Materials Society; 1995. p. 93.
[13] Bloyer DR, Venkateswara Rao KT, Ritchie RO. Metall Mater Trans A 1998;29:2483.